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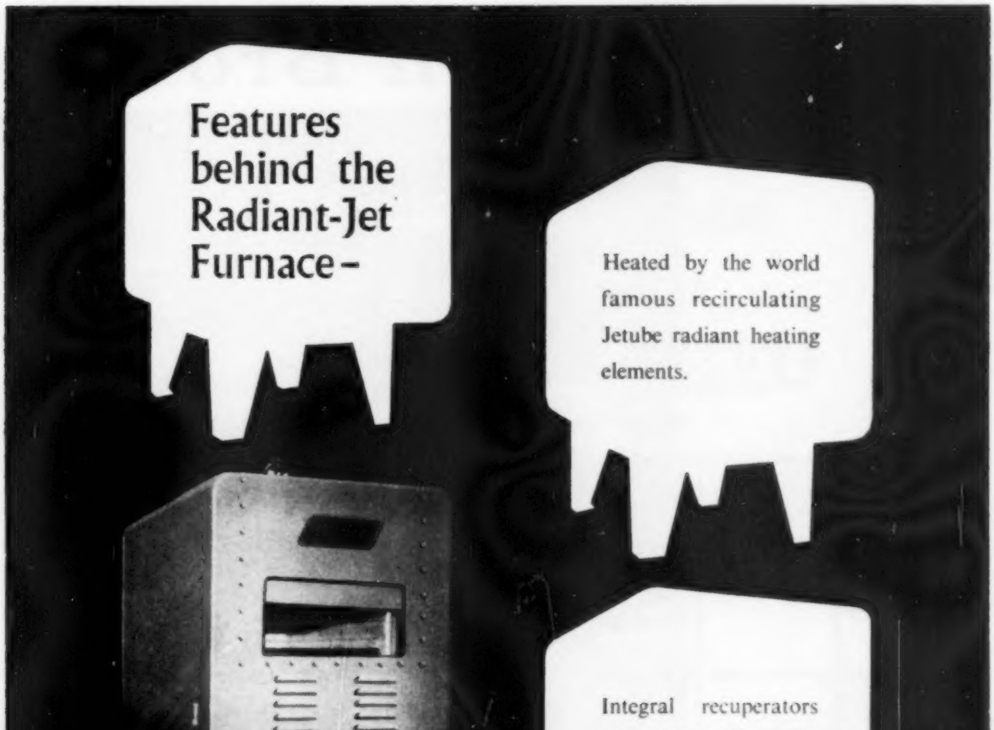
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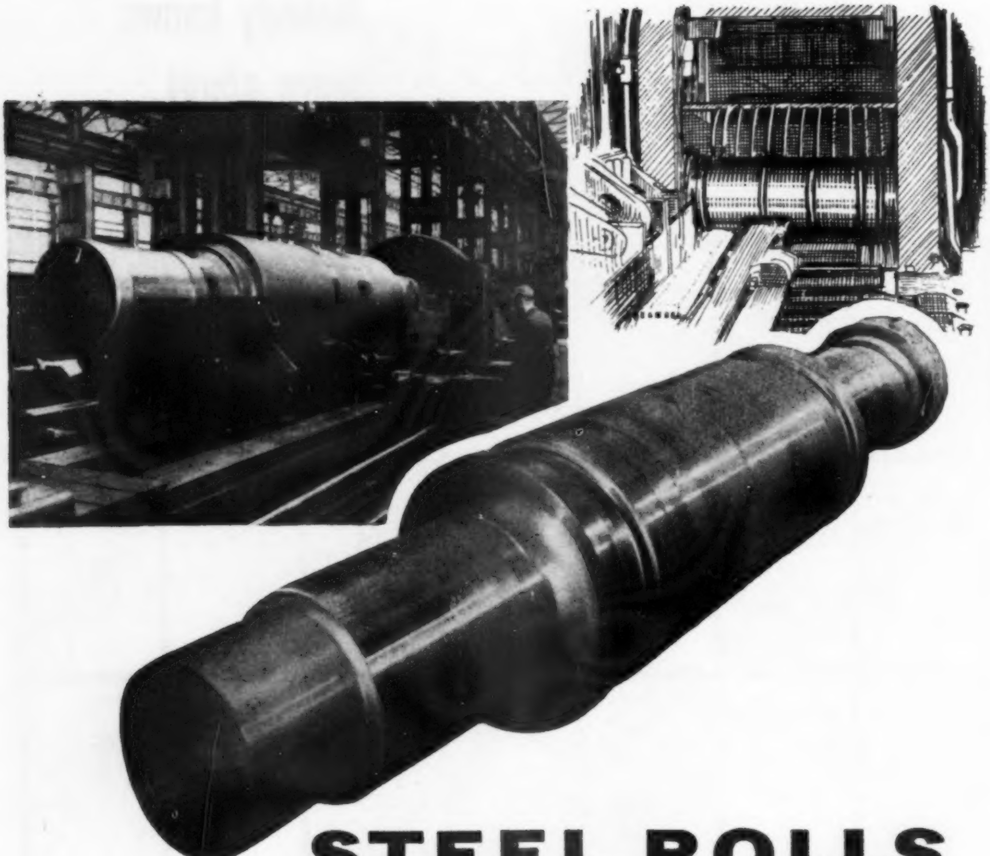

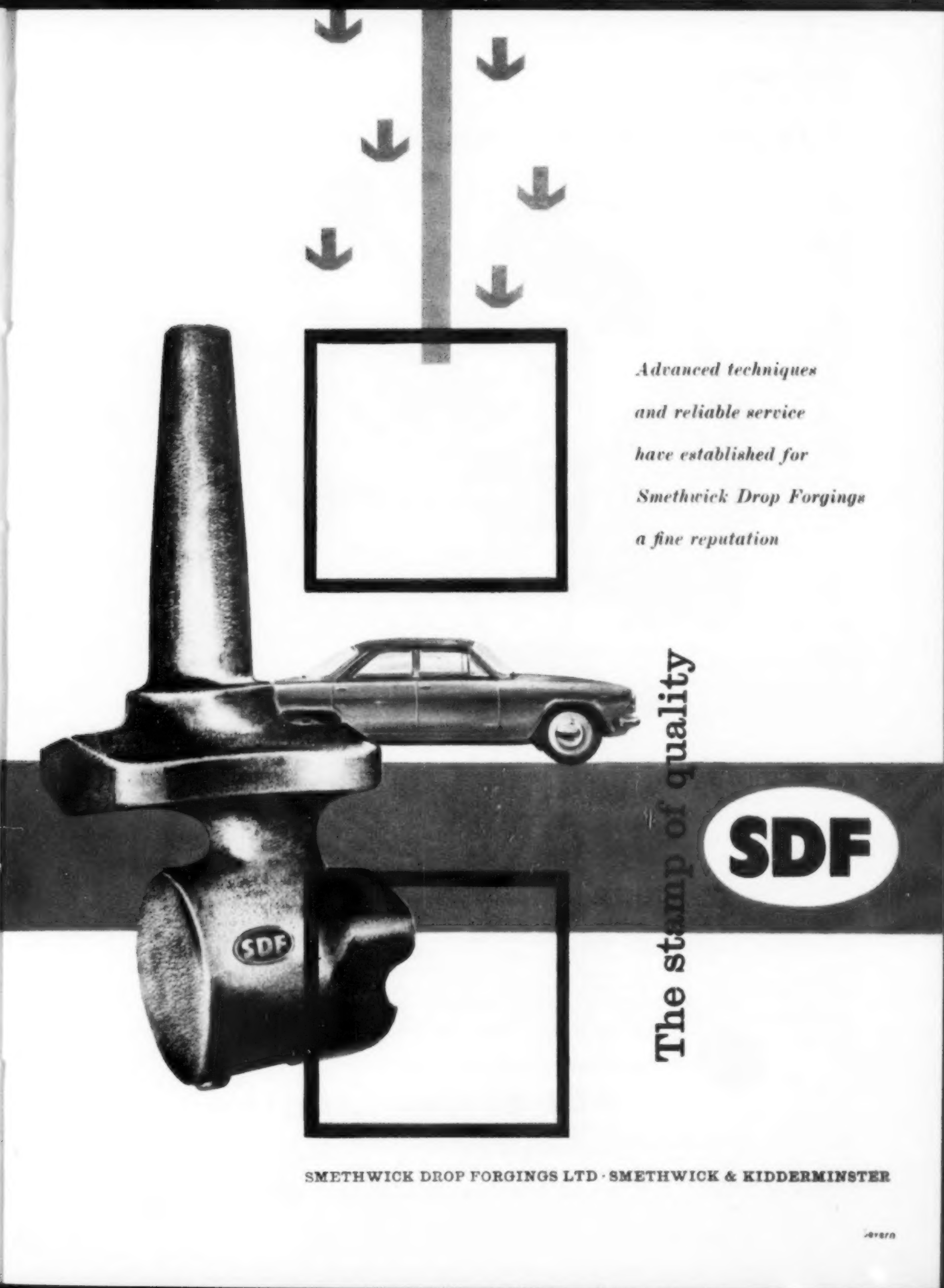
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Britain's exports

NATIONAL economic problems are hardly the direct concern of a technical journal but there are times when they cannot be wholly ignored. Technical developments have little meaning except in the context of their possible utilization in industry and, similarly, industry itself only makes sense in so far as it contributes to the real wealth of the nation. That, in fact, the contribution has been inadequate for many years to meet this country's expenditure has been clearly indicated by the relentlessly widening import-export gap. This does not represent any new state of affairs, any more than does the necessity of considering the problems of the Common Market—problems which might have been raised years ago. Now, nearly every day, we hear the slogan propounded that Britain must export more. The triteness of the proposed remedy makes it fatally easy to overlook the very real questions which it raises. It is one thing to make vague exhortations, and quite another to decide just who should export more and how the co-operation of the firms who have to do the work of exporting is to be enlisted. One expects the lead in a matter of national urgency to come from the Government, and the recent uninspiring Budget is hardly a cheerful omen.

As in most problems, the only valid approach is first to ascertain the facts. One pertinent field of enquiry is into the attitudes of individual companies to exporting, and it is interesting to find that preliminary research has recently been carried out by a private body and a report issued. The survey, 52 firms, was carried out by Marplan Ltd., on behalf of the Institute of Directors' Export Action Now Committee. The study was confined to firms having not more than 300 employees, since the problems in the export field of the giant enterprise are quite different from those facing the majority of British manufacturers. The method of analysis was essentially 'motivational'—that is to say, answers were not taken solely at their face value, but the techniques of social psychology were used to assess underlying attitudes.

The report considers that possibly the most pronounced feature of the enquiry was the difference in the amount of interest exhibited in exporting—and the conspicuously low proportion of managing directors who were really enthusiastic about, or even interested in, this side of their business. Many of the manufacturers tended to lack confidence in their ability to capture and sell in overseas markets. In a number of respects they felt their firms were too small to adapt their goods for foreign markets, or to employ an export manager to deal efficiently with their export trade, or to allow the managing director to travel abroad and make the necessary contacts. A fair number of the manufacturers interviewed were diffident about the success of the export drive, and looked on competitors such as Germany not with any idea of emulation but with resignation at their seizure of former British markets. Their attitude to new overseas markets was not adventurous, and many enquiries which come to the firm were ignored, or summarily dismissed as impossible.

Of the firms surveyed about 25% were exporting vigorously; the exports of 50% were declining or stationary at a low level; and just under 25% had never exported and were not really interested in doing so. It is not suggested that many definite conclusions can be drawn from the results of a survey of relatively few firms, but the figure of 75% for manufacturers not increasing, or not contemplating, exports suggests a general problem. A survey on a nation-wide scale might well prove a worth-while investment if it resulted in Government encouragement of the smaller companies with unrealized export potential. Even if the future of British exports does not lie with these firms, such encouragement might well succeed in stimulating a more vigorous attitude to production in general.

Precipitation of Laves phases in modified 12% Cr steels

DR. JAROSLAV KOUTSKÝ and DR. JAROSLAV JEŽEK

An analysis of the microstructures and X-ray analysis of specimens of modified 12% Cr steels have been carried out to study the precipitation of the Laves phases. The original publication of this work was in 'Hutnické Listy,' 1960, (11). The authors are at the Vitkovice Iron and Steel Works and the State Research Institute for Materials and Technology in Prague, respectively

In our earlier works,[1-5] we have published the results of a study of the structural stability of low-carbon, 12% chromium steels, alloyed with W, Mo, Co and V, intended for service at elevated temperatures. It was established that after long-term heating, especially at temperatures considered as operational, apart from the carbide phases in the structure, the Laves phases Fe_2W or Fe_2Mo are also in evidence. The chemical composition of the steels studied earlier is shown in table 1. With the exception of steels 1A and 4E, specimens of the remaining steels in the quenched and tempered state had a heterogeneous structure, formed by sorbite and a larger or smaller quantity of δ-ferrite. The precipitation of Laves phases in structurally heterogeneous steels has been observed primarily in the δ-ferrite, even though it is impossible to exclude their precipitation even in the sorbite.[3] In view of their similar morphology to that of the carbide phase present, in this instance the presence of a Laves phase can only be shown by means of the microscope with very great difficulty.

We have compared our results with the results found by Kuo, which he obtained during a study of the precipitates in Cr-W steels.[6] The chemical composition of steels with a medium and high W content, taken from this paper, is shown in table 2. According to Kuo, the decisive factor for the presence of a definite type of precipitate is not the absolute W content (or Mo content) or Cr content, but the atomic ratio W : C, Cr : C or W : Cr. If we calculate these ratios for the steels which we studied in former investigations (table 3), then our findings of the carbide phases present agree with the results of Kuo, as we have already partially mentioned earlier,[1] since, with the exception of

Table 1 *Chemical composition of the steel melts studied earlier, %*

Designation	C	Si	Mn	P	S	Cr	Ni	W	Mo	V	Co	N
1A	0·14	0·31	0·34	0·008	0·040	11·5	0·91	0·8	0·6	0·28	—	—
3B	0·13	0·39	0·38	0·006	0·035	11·94	0·18	3·9	0·66	0·24	—	0·02
3C	0·15	0·26	0·11	0·030	0·035	10·96	0·16	2·96	—	0·19	—	0·012
3D	0·10	0·25	0·12	0·030	0·032	11·44	0·16	3·95	—	—	—	0·062
3E	0·10	0·22	0·14	0·029	0·027	11·30	0·14	3·33	—	0·18	—	0·018
4A	0·21	0·08	0·96			11·10	0·10	—	2·6	0·25	—	
4E	0·21	0·06	1·48			12·9	0·48	3·6	0·16	0·30	5·6	

Table 2 *Chemical composition of, and carbides found in, the steels studied by Kuo,[6] %*

Designation	C	Cr	Ni	W	V	Atomic ratio W : C μ	Atomic ratio Cr : C *X*	Atomic ratio W : Cr *K*	Carbides found after heating at 700°C. for 125 h.
0351	0·32	1·15	3·41	5·30	—	1·1	0·88	1·25	$M_6C + M_{23}C_6$
0355	0·34	5·23	—	5·42	—	1·0	3·4	0·295	$M_6C + M_{23}C_6 + M_7C_3$
0359	0·30	8·72	—	5·12	—	1·1	6·6	0·167	$M_{23}C_6 + M_6C$
045 —13 ..	0·40	12·6	—	5·3	—	0·87	7·3	0·119	$M_{23}C_6$
0315— 5 ..	0·32	4·61	—	15·4	0·08	3·15	3·34	0·945	M_6C

steel 4A, in the quenched and tempered state and after long-term heating we established the presence of a single carbide phase—$M_{23}C_6$. At a high Cr : C ratio, above all the W : Cr ratio is decisive for the presence of the further carbide M_6C, and in all these steels the latter was less than 0·119 (table 2). Only in steel 4A was it higher (0·127, table 3), and in this steel, apart from the carbide $M_{23}C_6$, we also found the carbide M_6C.

The discrepancy between our results and those of Kuo rests in the fact that in no instance did Kuo find the intermetallic phase,[2] Fe_2W. It may be assumed that the possibility of its precipitation is decided primarily by the W : C ratio; in the steels studied by Kuo this ratio reached a maximum value of 3·15 (table 2), while in our steels it reached only 2·17 (table 3), if it is calculated from the average content of the elements present. In the quenched and tempered state the steels of Kuo had a clearly homogeneous, sorbitic structure, whereas in our steels 3B, 3C, 3D, 3E and 4A the structure in the quenched and tempered state was heterogeneous and, apart from sorbite, δ-ferrite was also present in the structure.

Hagel and Becht[7] investigated the structural stability of two types of modified 12% chromium steels in the range of temperatures from 480°–650°C. In a 12% Cr-Mo-W-V steel with an atomic ratio (W + Mo) : C = 0·975 and (W + Mo) : Cr = 0·064 after long-term heating in the range of temperatures indicated they found no other phase than the carbide $M_{23}C_6$. In the second steel—12% Cr-Co-W-V (Co content 4·5–5·5%)—at temperatures of about 485°C. they observed precipitation of ferritic particles enriched in chromium, and then at temperatures above 540°C. precipitation of the σ-phase. The atomic ratios Cr : C = 13·8, W : C = 0·975 and W : Cr = 0·071 are very close to the atomic ratios of our steel 4E. These conclusions were obtained both by a study of extraction replicas and electron beam diffraction of the precipitate included in them, and also by an X-ray study of an isolate obtained electrolytically. In the quenched and tempered state both the steels clearly had a purely sorbitic structure.

We have established the presence of Laves phases in 12% Cr steels alloyed with a greater quantity of W, and this has been confirmed by similar research work in the USSR. Yuganova and Nesterova[8] showed the presence of a Laves phase in steels EI 755 and EI 757. The chemical composition of these steels is shown in table 4. In the first steel the atomic ratio W : C equals 1·03, and in the second 1·73. In the quenched and tempered state in both steels a considerable quantity of δ-ferrite is present. It is true that the first steel has a lower W content, but this is compensated by the presence of the further, powerful, ferrite-forming element, Nb.

The low value of the overall atomic ratios W : C and Mo : C in our steels in connection with the existence of Laves phases has recently been indicated by Čadek.[9] In his opinion, the condition for the precipitation of Laves phases in molybdenum steels is a value of the Mo : C ratio above 5. In tungsten steels no Laves phase is precipitated, even where a W : C ratio of 3·3 is exceeded. Čadek tries to explain the presence of Laves phases in our steels primarily by the presence of V,

TABLE 3 *Values of the coefficients of the steels shown in Table 1*

Designation	Values of coefficients for overall chemical composition			Quantity of δ-ferrite, %, in the quenched and tempered state (determined metallographically)	Values of coefficients for phases: Ferritic			Values of coefficients for phases: Sorbitic (austenitic)			Precipitates found after heating at 650°C. for	
	Cr : C X	W : C (Mo) μ	W:Cr (Mo) K		Cr : C X	W : C (Mo) μ	W:Cr (Mo) K	Cr : C X	W:C (Mo) μ	W : Cr (Mo) K	1,500 h.	3,000 h.
1A	18·4	0·357	0·018	8	—	—	—	—	—	—	$M_{23}C_6$	$M_{23}C_6$
3B	21·1	1·95	0·071	48	94·5	11·1	0·12	12·75	0·93	0·073	$M_{23}C_6 + Fe_2W$	$M_{23}C_6 + Fe_2W$
3C	16·9	1·29	0·076	28	97·7	10·65	0·11	13·13	0·95	0·0727	$M_{23}C_6 + Fe_2W$	$M_{23}C_6 + Fe_2W$
3D	26·4	2·15	0·081	41	135·0	16·7	0·12	15·6	1·23	0·079	$M_{23}C_6 + Fe_2W$	$M_{23}C_6 + Fe_2W$
3E	26·2	2·17	0·083	43	133·0	13·7	0·103	16·0	1·02	0·064	$M_{23}C_6 + Fe_2W$	$M_{23}C_6 + Fe_2W$
4A	12·2	1·55	0·127	35	74·0	12·7	0·172	8·4	0·87	0·104	$M_{23}C_6 + Fe_2Mo + M_6C$	$M_{23}C_6 + M_6C + Fe_2Mo$
4E	14·2	1·12	0·079	0	—	—	—	—	—	—	$M_{23}C_6 + Fe_2W$	$M_{23}C_6 + Fe_2W$

TABLE 4 *Chemical composition of the steels studied in the research of Yuganova and Nesterova,[8] %*

Designation	C	Si	Mn	P	S	Cr	Ni	W	Mo	V	Nb
EI 755	0·13	0·38	0·79	0·012	0·018	10·85	0·31	2·05	0·73	0·09	0·37
EI 757	0·14	0·29	0·77	0·013	0·018	10·85	0·19	3·92	0·74	0·11	—

although he, of course, admits the possibility of the effect of heterogeneity of the structure on the equilibrium relationships.

Our opinion is somewhat different. We consider that even at these low overall values of the atomic ratios, in our steels with a heterogeneous structure the presence of vanadium is not an indispensable condition of the precipitation of Laves phases, since the atomic ratio W (or Mo) : C in the ferritic phase is substantially greater than the ratios given by Čadek as critical (table 3). The concentrations of the elements were calculated by means of the equations of Andrews, which we have described in our earlier papers.[1, 4]

We have already shown the presence of the Laves phase Fe_2W in steel 3D without the presence of V. We consider that, even in steel containing Co, which has a homogeneous structure in the quenched and tempered state, the presence of V is not a necessary condition for the precipitation of a Laves phase, but that such a condition is probably met by Co, which acts as a catalyst for the precipitation of Fe_2W at a lower W : C ratio than the critical. To verify our two opinions we conducted two melts, the first 12% Cr-Mo and the second 12% Cr-Co-W, the chemical compositions of which are given in table 5.

Results of the study of melts M and C

Steels M and C were melted in a 40-kg.-capacity induction furnace. After teeming, the ingots were soaked and forged into bars 14 mm. in diameter and 14 mm. long. These bars were quenched and tempered in accordance with table 6 and heat treated at temperatures of 650, 700 and 800°C. We investigated the microstructure of the specimens in the optical and electron microscope, and the qualitative composition of the precipitate obtained by chemical isolation in a 10% solution of bromine in methyl-alcohol. From certain specimens we also obtained an X-ray diagram of the precipitate included in an extraction replica. For the X-ray diffraction we used Cr_{K_α} radiation without a monochromator. The results of the X-ray analysis of the isolates are shown in table 6.

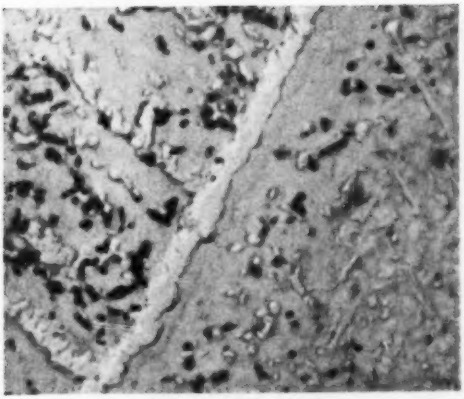
In steel M, there is about 25% δ-ferrite in the quenched and tempered state, and the Mo + C atomic ratio in this phase equals 9·5, which is a value relatively close to that of steel 4A. Even on heating at a temperature of 650°C. the intermetallic phase Fe_2Mo is apparent, and rapidly coalesces (fig. 1). As distinct from steel 4A, however, we find its existence even on heating at a temperature of 700°C. and, in fact, both by X-rays and also in the optical and the electron microscope (fig. 2). Only on reaching a temperature of 800°C. is the δ-ferrite devoid of any sort of precipitated particles (fig. 3), and on the X-ray diagram we also found no Fe_2Mo lines. The intermetallic phase Fe_2W we found already at a temperature of 650°C., and will describe it as the probably fine, and for the most part rod-like particles precipitated within the originally martensitic grains (fig. 4). The microstructure of steel C in the quenched and

TABLE 5 *Chemical composition and values of the coefficients of the melts studied*

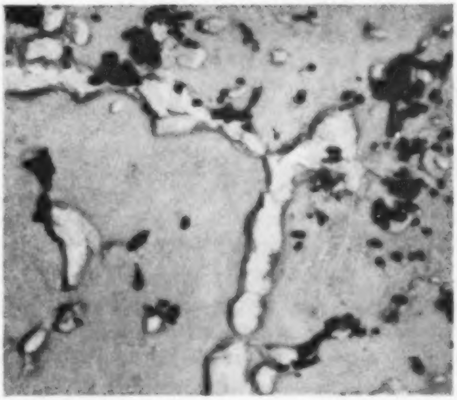
Designation	C	Si	Mn	P	S	Cr	Ni	W	Mo	Co	Values of the coefficients for the overall chemical composition			Quantity of δ-ferrite in the quenched and tempered state,	Values of the coefficients for the phases					
															Ferritic			Austenitic		
	%	%	%	%	%	%	%	%	%	%	X	μ	K	%	X	μ	K	X	μ	K
M	0·20	0·47	0·48	0·018	0·030	11·9	0·42	—	1·84	—	13·7	1·15	0·084	25	81·5	9·5	0·117	10·7	0·78	0·073
C	0·28	0·25	0·16	0·010	0·029	11·56	0·17	3·42	—	5·50	9·5	0·8	0·084	—	—	—	—	—	—	—

TABLE 6 *Results of the X-ray analysis of the isolates*

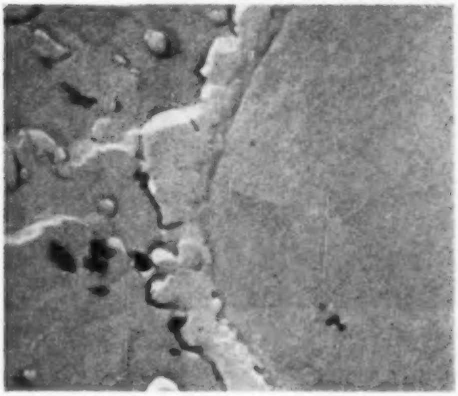
Steel	Heat treatment									
		Quenched and tempered as in the preceding column and heat treated								
	1,050°C./oil—770°C./1 h./air	650°C.			700°C.			800°C.		
		100 h.	500 h.	1,500 h.	100 h.	500 h.	1,500 h.	100 h.	500 h.	1,500 h.
M	$M_{23}C_6$ Cr_2N M_6C	$M_{23}C_6$ Fe_2Mo	$M_{23}C_6$ Fe_2Mo	—	$M_{23}C_6$ Fe_2Mo M_6C	$M_{23}C_6$ Fe_2Mo M_6C		$M_{23}C_6$ M_6C		
	1,050°C./oil—670°C./1 h./air									
C	$M_{23}C_6$ Cr_2N		$M_{23}C_6$ Cr_2N Fe_2W traces	$M_{23}C_6$ Cr_2N Fe_2W	$M_{23}C_6$ Cr_2N Fe_2W weak		$M_{23}C_6$ Fe_2W	$M_{23}C_6$ Fe_2W		$M_{23}C_6$ Fe_2W Cr_2N



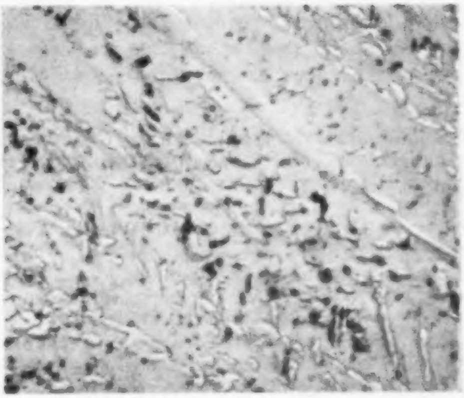
1 *Steel M, 650°C./100 h./air*



2 *Steel M, 700°C./100 h./air*



3 *Steel M, 800°C./100 h./air*



4 *Steel C, 650°C./500 h./air*

1-4 *Electron micrographs of extraction replicas shadowed with chromium. All specimens quenched and tempered: 1,050°C./oil + 770°C./1 h./air Subsequent heat treatment as shown under each figure* × 10,000

tempered state is purely sorbitic as that of steel 4E, and the values of its atomic ratios are also close to those of steel 4E. Just as in steel 4E, Fe_2W is present even at a temperature of 800°C.

In this instance also our opinions are basically valid, apart from the small variation indicated, that the phase Fe_2Mo is less stable than the phase Fe_2W.

Conclusions

By analysis of the microstructures and by X-ray analysis of the specimens of melts M and C it has been shown that:

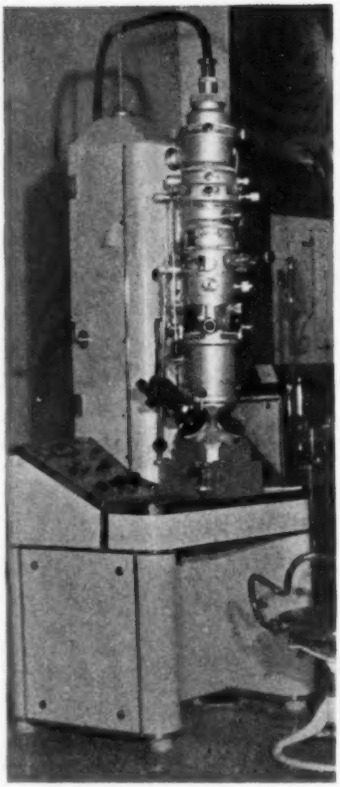
(1) Precipitation of the Laves phase Fe_2Mo can occur in 12% Cr-Mo steels with a low overall Mo : C atomic ratio, whose microstructure in the quenched and tempered state is heterogeneous, even when V is not present. This finding speaks in favour of our opinion that in these instances the differences in the concentration of the alloying elements between the ferritic and austenitic (sorbitic) phases have decisive importance.

(2) Precipitation of the Laves phase Fe_2W can take place in 12% Cr-W steels alloyed with a greater quantity of Co, which have a homogeneous sorbitic

continued on page 307

Some Russian electron microscopes

At the recent Soviet Exhibition in London, a range of Russian electron microscopes were on show. The specifications of these instruments do not suggest that the Russians have gained the lead in this field. Indeed, comparing instruments of similar size, higher powers of resolution seem to have been achieved by the best European manufacturers. Comparisons are difficult without a knowledge of how much these instruments would cost outside the Soviet Union, but it seems that some models are going to be made available in France. In view of this, it may be of interest to workers in the field to have at hand some details of this equipment.



The UEMV-100 universal electron microscope demonstrated at the Soviet Exhibition at Earls Court

THE HIGH-RESOLVING type EM5 electron microscope is designed for observation and photography of images of specimens with light and dark fields. This microscope may be also used for micro-diffraction study of separate portions of the specimens within 1 to 2μ by the micro-diffraction method, for stereoscopic photography of the specimens, as well as for electronographic examination in both transmitted and reflected electron rays.

The microscope is provided with an electromagnetic four-lens system, with the aid of which a magnified image of the specimen is formed on a fluorescent screen or photographic plate.

The condenser lens operates under reducing condition of 2·5 : 1, which allows to obtain with only one lens a diameter of the area lit by electrons of 10—8μ. The objective lens has a stigmator.

For obtaining comparison electronograms the design of the camera provides for the possibility of separate exposures of the two halves of the photographic plate.

Image focusing is performed by an optical binocular microscope with a magnification of $\times$8·5.

The electron gun of the microscope has an adjustable negative bias and an armoured high-voltage lead-in.

For preparing specimens for observation under the electron microscope, the equipment includes an electron-microscope vacuum system type EVP2.

The high output of the oil steam pump (200 l./sec.) and short length and large cross-section of vacuum piping make it possible to create the operating vacuum within 4–5 min.

SPECIFICATION

Resolving power	20Å
Electron-optical continuously variable magnification .. .. from	$\times$1,000–90,000
Size of photographic plate	45 $\times$ 60 mm.
Number of photographic plates in the magazine	4
Accelerating voltage	40, 50 and 60 kV.
Operating vacuum in the microscope column	1×10^{-4}–5×10^{-4} mm. of mercury
Mains supply voltage with neutral tap	220–380 V.
A.c. supply frequency	50 c./s.
Power consumption	2 kW.
Weight of microscope (without spare parts and accessories and without packing) .. .. approx.	850 kg.
Weight of the vacuum system without packing) .. .. approx.	300 kg.

Another model, the EM7, has a resolving power higher than 15 Å and a magnification range from $\times$1,000–150,000. Accelerating voltages are 50, 65 and 80 kV.

Universal model

The UEMV-100 universal electron microscope is a stationary laboratory instrument designed for visual and photographic study. The microscope permits study of: (1) objects in transmitted rays over a wide range of magnification with obtaining three-dimensional photographs; (2) objects in dark field; (3) electrograms of objects in transmitted rays; (4) microdiffraction images; (5) objects in incident rays; and (6) electrograms of objects in incident rays.

The microscope electron-optical system consists of electron gun, two condenser lenses, foscusing corrector and three lenses (objective, intermediate and projection), which provide for over-optical magnifications in the microscope. The objective lens as well as second condenser and intermediate lenses are fitted with stigmators. The intermediate lens provides for a gradual change in magnification, the image field remaining constant. Photographing the image is effected using photographic plates.

All the microscope's assemblies are fed from a three-phase a.c. mains through electron high-voltage regulators and lenses currents. Unregulation of high-voltage amounts to 0·0035%, and those of the objective lens current 0·001%, continuity of regulation being equal to 0·002%.

The microscope's vacuum system consists of a rotary initial vacuum pump and two oil pumps (booster and diffusion ones). The booster pump provides for the possibility of a protracted operation with the rotary pump being cut off. The vacuum system is fitted with traps to prevent oil vapours from penetrating into the microscope column from the pumps.

The object sluice has intermediate evacuation of the sluice portion of air. The photographic plates are changed when sluicing takes place.

SPECIFICATION

Acceleration voltage	100, 75 and 50 kV.
Beam current	5–200 μA
Minimum dia. of the beam on object	2–4 μ
Resolution	10Å
Magnification	×300–200,000
Tilt of object in 3-d photography ..	5 deg.
Maximum angle of diffraction ..	6 deg.
Object observation angle in incident rays	4 deg.
Binocular microscope magnification	×6
Working vacuum	3×10^{-4}–5×10^{-4} mm. Hg
Time of evacuation up to the working vacuum	25–30 min.
Permissible time of operation with the rotary pump being cut off ..	4 h.
Time of evacuation after the change of object	10–15 sec.
Size of photographic plates ..	6·5 × 9 cm.
Number of photographic plates ..	12
Object grid (diaphragm) dia. ..	2 mm.
Power consumption	4 kW.
Net weight	1,500 kg.

Electrostatic microscope

The MESM-40 small-size electrostatic electron microscope of simplified design is used for investigating specimens in transmitted rays as well as for obtaining stereoscopic images. The lenses unit being removed, the instrument may be used as an electronograph (electron diffraction camera).

The MESM-40 is designed in the form of two individual units. One of the units comprises microscope's column, feeding device and control panel, the other—vacuum system with a thermal spray installation and a spot welding attachment.

Vacuum system unit may be supplied separately for use as an installation for specimen preparing.

SPECIFICATION

Acceleration voltage	40–50 kV.
Resolving power	50Å
Magnification (four stages) ..	×1,500–10,000
Objective tilt angle in stereophotographing	4 deg.
Maximum angle of diffraction ..	8 deg.
Working vacuum	5×10^{-4} mm. Hg
Time of evacuation up to the operating vacuum	15–20 min.
Number of specimens charged in microscope	5
Size of photographic plates	4·5 × 6 cm.
Number of photographic plates ..	10
Specimen grid (diaphragm) dia. ..	3 mm.
Power consumption	1 kW.
Microscope overall dimensions ..	95 × 75 × 140 cm.
Net weight	200 kg.

Precipitation of Laves phases in modified 12% Cr steels

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structure in the quenched and tempered state, but a low overall W : C atomic ratio, even when V is not present. This finding confirms our opinion that in this instance it may be Co which acts as a catalyst for the precipitation of the intermetallic phase Fe_2W.

Acknowledgments

We consider it our pleasant task to thank J. Neid for collaboration with the X-ray structural analysis, and P. Schier of the Metallurgical Institute of the Czechoslovak Academy of Sciences for enabling us to carry out work on the electron microscope.

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New alloy steel from Samuel Fox

SAMUEL FOX AND COMPANY LTD., a subsidiary of the United Steel Companies Ltd., have started commercial production of a new alloy steel designed for service temperatures up to 675°C. Known as Esshete 1250, this austenitic creep-resisting steel is the outcome of five years of research and testing, with the result that unusually comprehensive data is available on its properties and performance.

Esshete 1250 is a 15% chromium, 10% nickel, 6% manganese steel, the composition including smaller percentages of silicon, molybdenum, vanadium, niobium and boron. It combines a high level of rupture strength with adequate ductility, good weldability, structural stability and oxidation resistance at elevated temperatures for long periods. It also has good manipulation properties. The steel is thus admirably suited for service in power stations of advanced design.

Esshete 1250 has proved satisfactory in the production of bars, tubes, pipes and large forgings; sheet trials are in progress. Large bars, pipes and plate can be readily fabricated in either the hot or cold condition, cold forming being understood to include temperatures not exceeding 650°C. and hot forming, not less than 900°C. Whichever method of fabrication is adopted, the material requires a re-solution treatment within the range 1,050/1,150°C.

In extensive weldability trials, employing various experimental and commercial electrodes, Esshete 1250 gave excellent results.

The structural stability of the steel compares most favourably with other more conventional austenitic steels. Its oxidation resistance has been assessed by a combination of laboratory tests and exposure in an experimental element in a power station for a period of approximately two years. In this case, too, Esshete 1250 has shown similar characteristics compared with other austenitic steels of the same chromium content.

The mechanical properties of the steel are shown in table 1.

Particular attention has been devoted to the stress-to-rupture properties of Esshete 1250, the test programme having covered a period of 25,000 h. to date. The results of this extensive testing are summarized in the following table 2.

TABLE 2 *Stress-to-rupture data (ton/sq. in.)*

Time (hours)	600°C.	625°C.	650°C.	675°C.	700°C.
100	24·0	20·5	18·0	16·3	15·5
300	22·0	18·5	10·5	15·1	14·6
1,000	20·0	17·0	15·0	13·5	13·0
3,000	18·5	16·0	14·0	12·0	10·0
10,000	16·8	14·5	12·5	10·4	7·5
30,000	(15·0)	(13·3)	(11·4)	(8·9)	(5·8)
100,000	(13·9)	(12·5)	(10·8)	(8·2)	(4·5)

Creep testing up to 10,000 hours has been carried out to determine the stress to give 0·1 and 0·2% strain. The results of these tests are summarized in the table 3.

TABLE 3 *Creep-test data*

°C.	Stress for 0·1% creep strain (ton/sq. in.)			Stress for 0·2% creep strain (ton/sq. in.)		
	1,000 h.	3,000 h.	10,000 h.	1,000 h.	3,000 h.	10,000 h.
650	7·60	6·00	5·15	7·80	7·25	6·65
675	6·55	5·45	4·70	7·30	6·50	5·60
700	4·90	3·60	1·80	5·60	5·30	3·10

Like other austenitic steels in the solution-treated condition, Esshete 1250 has low proof-stress values. These can be improved by warm working, and this process is found, in fact, to push up the 0·1% proof stress to 32 ton/sq. in. at room temperature and 23 ton/sq. in. at 625°C., while still holding elongations on $4\sqrt{A}$ of the order of 48 and 25% respectively. The hardness of the steel is not altered at 625°C. but at room temperature is the equivalent to 43 ton/sq. in. U.T.S., which is only slightly more than in the normally rolled condition.

Warm working also improves the creep properties, so that the steel is well suited in this condition for use as a high-temperature bolting material.

TABLE 1 *Typical mechanical properties—solution treated*

Test temperature (°C.)	Tensile strength (ton/sq. in.)	Proof stress (ton/sq. in.) (0·1%)	Proof stress (ton/sq. in.) (0·2%)	Limit of proportionality (ton/sq. in.)	Elongation on $4\sqrt{A}$(%)	Reduction of area (%)	Modulus of elasticity (ton/sq. in.)
20	39·0	13·8	15·0	7·0	67·0	69·6	13,400
100	34·3	10·6	11·6	5·2	64·5	72·0	13,000
200	31·8	9·9	10·1	4·2	55·0	70·3	12,500
300	31·4	8·5	9·4	3·6	45·0	64·0	11,900
400	31·5	8·6	9·1	4·4	47·7	60·8	11,300
500	30·3	8·2	8·8	4·8	40·5	60·8	11,700
600	28·5	8·2	8·6	4·8	47·4	64·0	10,200
700	23·0	8·1	8·5	3·6	40·0	51·2	9,650

Metallurgical aspects of the cold extrusion of steel

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The paper considers the effect of extrusion on the metallic materials involved at various stages of the process, the ways of bringing these materials to a condition suitable for working, and the effects of the extrusion process on later treatments. Side by side with these conditions the factors defining the practical limitations are indicated. The author is with the Cold Extrusion Division of Forgings and Presswork Ltd., Birmingham*

A LOGICAL BEGINNING to the subject can be made by considering the material which is being pressed. Metal which is subjected to cold-forging processes must withstand heavy plastic working resulting in bodily displacement. Internal movement often exceeds the obvious external change of shape. The metal exerts a resistance to the deforming forces, and the resistance increases in a well-known manner with the degree of working. A convenient measure of the increase is the hardness of the metal or its tensile strength. If either of these parameters is plotted against deformation a curve is obtained such as that of fig. 1. The practical measure of the work-hardening phenomenon is the pressure exerted on the tool surfaces which must not exceed the load capable of being borne by the tool material. Hence the first boundary for extrusion possibilities is quickly reached, since an alloy whose resistance to deformation exceeds the long-term load-bearing capabilities of the tool steel cannot be extruded under production conditions.

The plastic properties of the metal itself have to be taken into account, for a suitable material must be sufficiently ductile to withstand heavy reduction without rupture, and the increase in strength brought about by work hardening must be only gradual. A work-hardening curve showing a slow increase in hardness with deformation characterizes the steels most suited to cold forming. It is possible to take an alloy which subjects the tools to a reasonable loading at the beginning of the press stroke but, as the base is approached in a backward extrusion for example, the load can rise above that permissible.

Both work-hardening properties, and the extent to which a steel allows itself to be deformed without rupture, are closely related to chemical composition for a given structure. Increases in carbon and other alloy contents lead automatically to increases in resistance to deformation and since extrusion is limited by the tools, alloys must be limited to those which will not exert reactive pressures greater than the tools can bear. Prior thermal treatment has a great bearing on the question but, for this comparison, it is assumed that the best plastic condition has been achieved. For maximum deformation a metal should contain the minimum alloying elements including impurities. On this basis, the most suitable steel is a plain carbon steel containing less than 0·1% carbon, and such a steel will submit to reductions in area of the order of 90% under suitable conditions, without overstressing the tools. No practical steel can be guaranteed to continue sufficiently uniform to allow such a high reduction with safety, and figures as high as this cannot be assumed in practice. As the complexity of the steel increases, the work which can be conducted in one operation becomes less and, as a consequence, the number of intermediate anneals and surface treatments rises. A third limit to composition is then reached, for the increased costs do not allow such an alloy to be worked economically.

It can be more specifically stated that low-carbon steels are capable of being drastically deformed and may be made into complex shapes. The extent of reduction has to be taken more

* Article based by the author on his lecture given at the Wolverhampton and Staffordshire College of Technology last March at a two-day symposium on 'Cold flow forming.' A further article based on the lecture by Mr. R. A. P. Morgan, O.B.E., will be appearing shortly in METAL TREATMENT.

carefully into account when the carbon exceeds 0·2% and depths of penetration have to be watched. Practical extrusion ceases above about 0·35% carbon. Metallic elements exert their effects superimposed upon that of carbon in any one type of steel.

For an appreciation of the effects of individual elements it is helpful to consider what happens when a piece of steel is cold worked. The structure can be assumed to be either ferrite/pearlite or free carbide in ferrite, depending upon composition and previous annealing treatments. The ferrite constituent of an unalloyed steel is soft and most of the deformation takes place by plastic movement of these crystals of solid solution. Carbide in either of the two stated forms serves to hinder flow, but the matrix of ferrite is very plastic and is able to accommodate the hindrance.

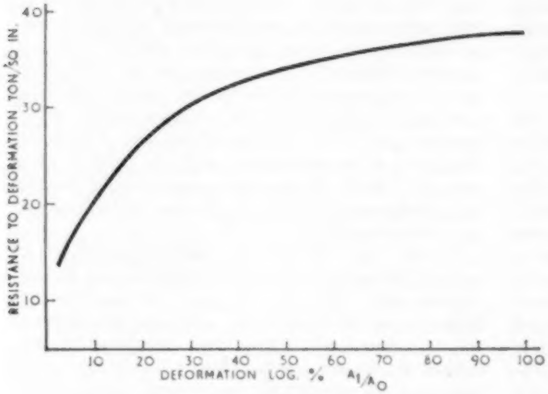
Following this picture a little further it can be seen that the effect of individual elements is dependent upon the way in which they distribute themselves among the constituents of the structure. The carbide is in any case hard and brittle and can therefore take no part in plastic movement. If, then, the additional element can be induced to confine itself to forming an alloy carbide, the effect of that element will be small, *i.e.* it is to all intents and purposes removed from the plastic system. In the case of an alloy which persists in solid solution in the ferrite, there is a direct unfavourable effect on the most important part of the system from the plasticity viewpoint, and the properties suffer. It follows that strong carbide-forming elements such as chromium do not have such a deleterious effect on plastic properties as ferrite solution elements. Elements which divide themselves between the constituents, *e.g.* manganese, are naturally intermediate in their effect on the behaviour of a steel on cold working. At the present state of knowledge, large-scale extensive cold forming can be applied to steels containing as an upper limit about 3/4% nickel, 1½% chromium, 1½% manganese and ½% molybdenum. Unfortunately, British carburizing steels are not of the manganese-chromium type which are popular on the Continent, and which lend themselves to extrusion methods.

In the laboratory the plastic properties of a given steel can be determined by upsetting a cylindrical test-piece between cone-faced platens. The object of the conical facings is to exert an outward radial thrust, thus overcoming the friction forces which hold the top and bottom faces of a plain cylindrical specimen and cause it to assume a barrel shape. In the test, the cylinder is upset in known increments and the corresponding load on the press indicator is taken and converted to stress by dividing by the cross-sectional area at each increment. The degree of area increase is then calculated as a measure of deformation. Fig. 1 is an example of a curve plotted from information derived in this manner. The test is due to Siebel and Pomp.

Having established that steel for cold extrusion must be capable of sustaining heavy internal shearing without rupture, by which is meant that individual layers must be capable of moving over other layers which are relatively stationary, the forms of steel which are available for use on an industrial scale can be discussed. In general, interfaces between constituents of differing plastic properties should be reduced to a minimum. Such interfaces exist around segregates and inclusions for example. Hence elements forming non-metallic inclusions such as sulphur and phosphorus must be held to as low a limit as practicable. Such a condition of high-inclusion content is precisely the aim of the producer of free-cutting steels and, for this reason, such steels cannot be subjected to drastic cold working. In a similar manner the rim of pure iron on rimming-quality steels represents the junction between layers of differing plasticity. Layers in rimming steels have been known to part under the conditions of shear existing in cold-forming processes. This phenomenon occurs under certain conditions even with sound uniform steels and, in fact, imposes a lower limit on the degree of deformation which can be undertaken on backward extrusion.

Under these conditions, fully killed steel is chosen for extrusion purposes in spite of the associated casting difficulties, which result in low ingot yield and surface imperfections. In the U.K. further considerations dictate that the steel must be of open-hearth or electric quality as the high nitrogen content of air-blown steels makes the steel very prone to strain ageing. Strain-age embrittlement is to be avoided with extruded parts, and excess aluminium is added to the steel to avoid the possibility of a heavily worked piece becoming brittle with time. The exact percentage of residual aluminium which must be present is in some dispute among different authorities. Some maintain that 0·04% is essential, whilst others claim 0·02% to be sufficient.

Independent of the steelmaking process, steel for extrusion must have the indicated desirable properties for practical extrusion to be successful. It is important to differentiate between trials and production as most steels are extrudable to a degree. Extrusion possibilities are bounded by the performance of tools and steel under working conditions. Annealing techniques which bring the material into the condition and structure most suited to working enable the boundaries to be extended into the working of more complex steels,



1 *Relationship between degree of deformation and resistance to deformation*

and the production of more complicated flow patterns. Once again, it is cost which determines whether more expensive annealing treatments can be used to ease tooling problems.

From the purely physical viewpoint it is desirable to have a uniform distribution of all constituents in the structure and to avoid grain-boundary precipitates as far as possible. A good finish on the surface of the steel is important since there is only a limited opportunity for removal of imperfections during processing. In the hot forge, surface scaling can remove slight seams and allow substandard material to be successfully forged. Black hot-rolled forging bar, nevertheless, has been found to be quite suitable for use if the quality is carefully controlled.

Thermal treatments

During the production of extruded parts, a number of thermal treatments are applied to the billets with the object of bringing them into a condition suitable for cold working or reworking. The complexity of the treatment depends on steel composition and occasionally on the degree of deformation which has been executed, or is about to be executed.

A billet of low-carbon fully-killed plain carbon steel allows itself to be heavily deformed after the simplest of softening treatments. It very occasionally may be necessary to submit rolled material to a normalizing treatment when abnormally large non-uniform grains are present. Usually, however, cooling from a subcritical temperature after a reasonable soaking period is sufficient for the severest deformations. Quite reasonable degrees of cold working can be conducted on hot-rolled bar in the condition in which it is received from the steel supplier.

There seems little to gain from attempting to spheroidize the carbide present in simple low-carbon steels as there is always a preponderance of soft plastic ferrite to carry the carbide or pearlite. When the carbon decreases to such a low level that the carbide no longer appears as pearlite, but forms as a separate constituent, it is preferable for it to appear as globules rather than as a grain-boundary film. The effect of such a film is not, however, so disturbing as it can be in deep drawing work where the main forming force is tensile and where grain boundary carbide can instigate intergranular cracking.

Once a steel has been cold worked, recrystallization takes place independently of allotropic changes, and steel behaves as any other metal in this respect. Interstage annealing of an extruded plain-carbon steel normally takes the form of a subcritical heating and cooling which brings about recrystallization. The slower the cooling, of course, the more complete the removal of dissolved carbon from the ferrite according to the well-known solubility curve of the iron–iron carbide diagram. The extent of completeness of precipitation is of no great practical significance, providing abnormally fast cooling is not employed, for at least the first 100°C. of cooling below the annealing temperature of around 650°C.

It is not unknown to discover that a certain location in an extruded part has developed abnormally large grains. The phenomenon is common to all cold-working processes, and is the familiar critical strain-grain growth. Between a certain range of degrees of strain, nucleation on annealing is very slow so that the number of crystals in a given volume is small with the result that individual crystals are very large. They can be so large that etched surfaces display grains readily discernible to the naked eye. Further working of such a structure is not easy as rupture occurs

readily and the surface of a strained large-grained specimen has the characteristic rough 'orange peel' appearance. A finished component with such grains is correspondingly weak, especially when subjected to shock loading.

Recrystallization is the only answer, and this can be brought about in two ways. The first is to cold work further, with a degree of strain above the critical range, which varies between 5 and 20% reduction, depending upon individual specimens. Such a treatment is fairly readily applied, for example, to strip, which can be re-rolled, but is not so easily applied to pressings from strip or cold extrusions. Fortunately, there is another way of inducing recrystallization by using an allotropic change to build a new set of crystals in another phase. Steel exhibits a very convenient $\alpha \rightarrow \gamma \rightarrow \alpha$ transition which enables a simple normalizing or high-temperature annealing treatment to bring the coarse-grained specimens back to their normal fine-grained structure suitable for use or reworking.

As soon as alloying elements are added to the plain carbon steels, punch loads for a given reduction increase, particularly when the added element persists in solid solution in the ferrite. The simple annealing cycles suitable for the carbon steels no longer suffice to bring the alloy steels into their most plastic conditions. The objects of a softening treatment must be: (*a*) to place as much of the alloy constituent as possible into carbide, and (*b*) to reduce the amount of all elements dissolved in the ferrite to as low a value as possible. Following these achievements the configuration of the metallographic constituents can be considered.

Both initial objectives are conditions which would be present if equilibrium could be attained in the system, and are achieved in practice by cooling from annealing temperature as slowly as possible. Some elements, notably nickel and copper, do not form carbides, and remain persistently in alpha solid solution. Whereas the small solid solubility of carbide below A_1 temperature had little effect in the case of carbon steels, the rate of cooling of the alloy steels below A_1 is found to have an appreciable effect on the plastic properties and the hardness.

Assuming that by a suitable cooling rate the alloy elements including carbon have been placed in the best possible form which their respective properties allow, the arrangement of the structure as a whole can be considered. Many authorities recommend that for highest plasticity the carbide should be spheroidized and uniformly distributed. In a medium- or high-carbon steel the structure is predominantly pearlite grains which are not very amenable to plastic working at room temperature. The ferrite which is capable of flow is largely interlamellar and as such is bounded and restricted by hard carbide plates. Such ferrite grains which do exist are locked by neighbouring pearlite grains. In such a case, it is clear that the only way to provide sufficient continuous ductile ferrite to sustain deformation is to divorce the carbide from the pearlite and transform it from plates which restrict flow into spheres which are completely enveloped in a soft matrix and are carried along with it. Such a structure undoubtedly enables high-carbon steels to be drawn into industrial toe-caps for example.

It might be expected that the more difficult of extrudable steels would also benefit from an annealing treatment which should place the carbides in a fully spheroidized condition, accepting always that a suitable slow cool follow that treatment to precipitate as much alloy as possible as carbide. The obtaining of uniformly distributed spheroidized carbide from hot-rolled or normalized stock necessitates a lengthy heat treatment even if thermal oscillation is employed to induce alternate solution and deposition. Spheroidization occurs during prolonged holding of the steel at as near to Ar_1 as can be managed under practical conditions. Experiments were carried out to attempt to speed the attainment of uniformity and spheroidization. Some success in speeding the rate of spheroidization can be achieved by cold working prior to annealing, which provides nucleation energy for the plates to re-form into spheroids. An overtempered martensite has a granular structure and is very uniform in its distribution. It was considered that a structure of this type presenting uniformly placed carbide nucleii might be a good starting point for spheroidization. The result of an annealing cycle on such a structure was a beautifully uniform spheroidized microstructure which was achieved in a fairly short time. The steel seemed to be in a condition ideally suited for cold working, but surprisingly, both the hardness value and the behaviour under the press were disappointing.

A reappraisement of the situation leads to the conclusion that the conditions governing the plastic working of high-carbon steels are essentially different from those relating to the working of low-carbon steels even if the latter contain alloying elements to increase their stiffness. In the case of high-carbon steels, most of the constituent capable of movement is locked by hard carbide, and must be freed by annealing and turned into a continuous matrix. With low-carbon steels there is plenty of ferrite, and the effect of freeing the interlamellar ferrite is insignificant. There is little, therefore, to be gained by spheroidization. In addition, if the carbide is distributed finely and uniformly the situation is worsened, and the system presents a structure less amenable to cold work than one consisting of large open blocks of almost

pure ferrite grains which can carry the pearlite.

In summing up the effects of annealing and constituents on the behaviour of a steel in cold extrusion, it is apparent that the important constituent must be the ferrite and annealing must have for its object the formation of a continuous open matrix of the softest possible ferrite containing the lowest concentration of alloy in solid solution which can be achieved in practice. The situation is achieved by annealing to obtain a ferrite/pearlite structure which is cooled as slowly as possible to allow maximum precipitation.

Outside of the purely metallurgical treatments there are other preparatory treatments which have as their objective the provision of a lubricant barrier between tools and extruded steel. The barrier prevents metal to metal contact and consequent seizing which would occur under the heavy extrusion pressures. The predominant lubricant is a crystalline phosphate system, and prior to its application it is usual to remove scale, and prepare the surface for phosphating by pickling in dilute acid. At this stage there can be a danger of introducing hydrogen into the steel. Surface cracking has been known to follow severe pick-up and precautions against hydrogen embrittlement are worth while.

Metal flow and subsequent effects

The metallurgical system existing prior to working or reworking has been discussed in some detail. Attention can now be given to the behaviour of the metal during working, and the properties which the cold working induces.

The flow of metal in a forging die can be far more complex than the change in external shape indicates. When the operation is conducted hot, the material is made reasonably uniform in its plastic behaviour and internal shearing takes place readily without rupture. There is also no complication ensuing from work hardening which has the effect of increasing resistance as flow proceeds. If an attempt is made to emulate hot-forging practice, it is quickly found that cold metal is no longer amenable to such complex flow, *i.e.* it is not merely a matter of doing the same things with more power. Flow patterns which are too complex can easily result in rupture between moving and stationary layers. Such cracking may or may not appear on the surface, and the metallurgist can here provide help to the extrusion engineer by examining flow patterns and suggesting how they might be simplified or by indicating danger zones. Some very interesting results have been published in the recent NEL reports on the effects of die angle on flow patterns for simple extruded shapes.

So far, the effect of work hardening has been discussed from the viewpoint of its hindrance to continued flow, and the effect on the tools. The property of work hardening can be usefully employed in that the increased strength of a worked metal can be of value. Owing to the heavy reductions involved in cold extrusion the properties induced are reasonably uniform throughout the section—certainly no worse than the uniformity of a heat-treated steel. There have been many descriptions of properties of extruded parts related at times to measurements of the reduction associated with these properties. Measurements based on sections cut from components can be misleading as the shape is no guarantee of deformation which has actually taken place. Further, there are few figures relating to the effect of extrusion on notched impact values which are regarded by many as being important.

A series of specimens were prepared in such a manner as to make the flow pattern as simple as possible, *i.e.* so that external shape could be used to measure deformation with but little error. The specimens were prepared by forward extrusion of solid rods and the final diameter in each case was 0·450 in. so that standard Izod round test-pieces could be produced without machining other than notching. Sub-standard tensile tests were also conducted on sections of the same samples so that all results could be correlated. Figs. 2, 3 and 4 illustrate the results of the tests in the form of curves.

Figs. 2 and 3 show very little difference in properties, which indicates that the aluminium addition is in fact suppressing strain-age effects. Fig. 4 shows the marked effect of a stress-relieving anneal which increases notched impact value tremendously without sacrificing much tensile strength. The combination of properties of 38 ton/sq. in. and 80 ft./lb. Izod value is remarkable.

A puzzling factor was the rise in impact value with increasing reduction. This trend is the reverse of that which would be expected. The same trend is noticed in all conditions examined and cannot therefore be due to a strain-ageing effect at low degrees of deformation. It was eventually discovered that the particular steel involved exhibited a continuous network of carbide around its grain boundaries and it was considered that this was the reason for the poor values. Extrusion altered the shape of the grain and diminished the effect of the carbide.

The figures show that the yield stress approaches the ultimate tensile stress very closely, and for this reason it is much more relevant to consider this figure when making comparisons between properties of extruded and heat-treated steels.

There are indications that the fatigue life is consistently better for an extruded part than its equivalent machined from bar or a forging. As yet,

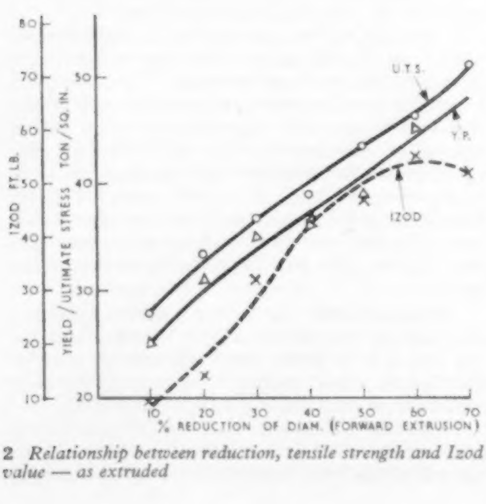
experimental evidence cannot be quoted, but it seems reasonable to assume that the excellent surface finish and grain flow, together with the system of compressive stresses at the surface, should have a beneficial effect. Owing to the small amount of machining which is left on an extrusion there is reduced likelihood of severing flow lines and exposing inclusions which might act as stress raisers under conditions of fatigue.

Effects of the extrusion process on subsequent treatments

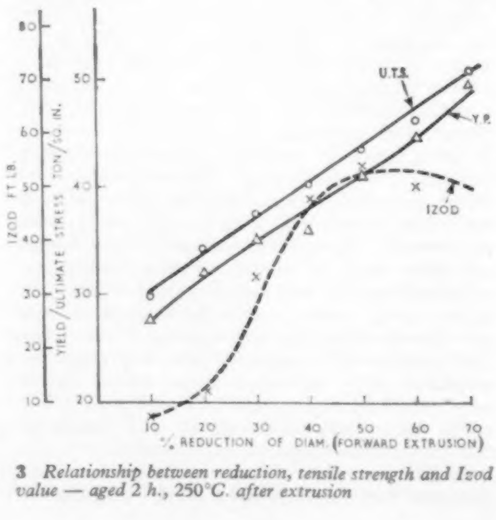
Once cold-deformed metals have been subjected to a thermal treatment which involves heating to above recrystallization temperature, the effect of the cold working is destroyed. It is to be expected, therefore, that such a treatment will exert its normal effects on cold-formed parts in a given steel, and this has been found to be true. It is in what might be termed the 'secondary effects' where the process of cold forging or extrusion might be expected to have influence. The phosphate coating could have a retarding effect on carbon acceptance in carburizing for example, or the residual stresses in an extruded part might cause distortion.

These questions can now be answered with authority based on experience gained with many thousands of parts in both carbon and low-alloy steels. The effect of the phosphate on rate of carburization has been found to be undetectable in practice. No undue distortion on carburizing cold extrusions has been reported. In the instances known where an extrusion has replaced a piece turned from bar or hot forging, consistently less movement during heat treatment has been reported. Feldmann suggests that the heating which occurs during extrusion may result in some relief of the stresses. The temperature normally attained is of the order of 200°C. It is also probable that the stresses are rather uniformly balanced so that their relief during heat treatment does not have such ill-effects as the relief of the stresses induced by heavy machining of substantial volumes of metal. A consequence of low distortion is that the normalizing treatment which often precedes carburizing can be omitted. The aims of such a treatment are to refine large grains resulting from hot work, and to allow distortion to take place before final machining. In the case of an extrusion, neither of these contingencies arises.

Welding is a process which is accompanied by thermal effects. In and around the weld zone a cold-worked steel will soften, the extent of the affected area depending upon the degree of spread of heat. The loss of strength must be borne in mind in design considerations. As a comparison the effects of welding on a component which owes its strength to heat treatment can be cited. In and near the weld zone the structure will be over-tempered and soft which corresponds to the softening of a cold-worked structure. At a certain location away from the weld there is often a zone which has been heated to a temperature above Ac_3 and quenched, when the heat source has been removed, by the surrounding mass of metal. A hard zone has been thus developed which in certain applications could lead to cracking and is for preference to be softened by a tempering treatment. The lack of response to quenching of a low-carbon steel does not permit the formation of such a hard



2 *Relationship between reduction, tensile strength and Izod value — as extruded*



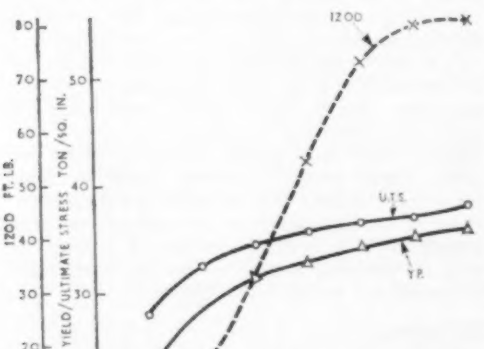
3 *Relationship between reduction, tensile strength and Izod value — aged 2 h., 250°C. after extrusion*

zone in a welded cold-worked steel, although as a counterbalance it must be said that the softening will be greater in the soft zone because of the low-carbon content.

The phosphate coating has certain properties which do have effects on after-treatments and uses. It possesses a certain power of protection against corrosion which helps to reduce deterioration in storage. It also has good lubricating properties and may help to reduce wear on hydraulic cylinders, for example, particularly during the running-in period. On the other hand, the phosphate has fairly good dielectric properties which hinder the application of electro-plated coatings. It also prevents metal-to-metal contact at quite high temperatures and hot-dip plating and allied fusion treatments such as soldering are difficult unless a particularly active flux is used. Welding is also adversely affected, and the coating should be removed before welding.

Suitable methods of phosphate removal can be both local or general. Mechanical and chemical methods are available. Clearly any mechanical cutting process, including wire-brushing and shot-blasting, can be applied. The phosphate is also soluble in both acid and alkali solutions.

It has been mentioned that free-cutting steels are not generally extrudable. The steel used for extrusion is best machined in the cold-worked condition, for the effect of the work hardening is to improve the machining properties of the otherwise soft steel. A parallel might here be drawn with the beneficial effects of bright drawing mild-steel bar. Machinability of an unannealed extrusion can be expected to be similar to that of a medium-carbon steel with a strength of about 40 ton/sq. in. Annealing of the extrusion will result in poorer machinability. These remarks apply, of course, to carbon steels.



4 *Relationship between reduction, tensile strength and Izod value — stress relieved 550°C., 2 h. after extrusion*

5 *Group of typical extrusions*

Defects arising in cold extrusion

As with any production process, the cold-extrusion method is not free from the possibility of defects. Failings in the original bar material are carried through to the finished part, although many defects are shown up during processing and result in early rejection. There are, however, certain defects which may be caused by conditions arising in the process itself. Most of them are due to excessive internal shear or to working the material beyond its capability to yield.

Cracks on the edges of the flanges and similar situations may be due to dirty steel or steel poor in ductility. Such cracking is due to the inability of the particular steel to extend under the tensile stress induced in the periphery by the expanding of the flange. Cracking of this type is a normal occurrence in many metalworking processes where a discontinuity exists in the metal and requires, therefore, no further explanation.

Internal cracks can result from a variety of causes. Care must be taken when the base of a

backward extruded cup becomes thinner than the thickness of the wall. This relationship is an empirical one based upon observation. The incidence of base cracking as the base is thinned is a consequence of changes in flow as the punch approaches the bottom of its stroke. At this stage the effective orifice between punch nose and die becomes restricted, and the zone of 'dead' metal immediately preceding the punch is forced to move.

A second source of internal cracking is the complication in flow pattern when the reduction on backward extrusion is too low. As the relationship of area of punch to area of cup wall decreases, *i.e.* the reduction decreases, the metal remote from the punch is affected increasingly less by the penetration, and in the extreme the outer section acts as a second die with all internal movement of metal confined to the zone near the punch. Consequently, at a certain stage, depending upon the particular metal, the internal shear becomes too great and actual rupture occurs. Internal fractures of this nature may escape notice until a second operation causes them to break the surface, and care must be taken to avoid the production of a heavy percentage of scrap under conditions conducive to the formation of this kind of crack.

Finally there are the defects which can be attributed to the presence of oil or water on the steel during extrusion. The surface rippling effect, which is quite familiar in the cold working of aluminium with excess lubricant, has its parallel in steel when oil or water is allowed to enter the system. Surface damage can be extended by following operations to give rise to folds and tears of sizable proportion. The effect is most commonly found on internal surfaces and the discovery of the results can be very disturbing until the cause is known.

Acknowledgments

Acknowledgment is given to Forgings and Presswork Ltd. for opportunity to prepare this paper.

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Avoidance of distortion in heat treatment

concluded from page 323

even in a symmetrical component), or of segregation, which is bound to be fairly randomly distributed.

Another important source of asymmetric distortion can operate in tool and die steels with carbide stringers, causing anisotropic expansion behaviour. Anisotropy is more severe the coarser the carbide particles.[3] The directionality of the carbides is influenced by the degree and method of forging, the particle size depends mainly on the solidification of the ingot.[6]

The influence of these differences in carbide distribution may be sufficient to cause severe warping even in steels with enough retained austenite to balance the expansion due to martensite formation. It is not surprising that it cannot be completely cured. The effect is less severe in air hardening than in oil hardening steel. Milder quenching, perhaps even faster heating, might help to counteract the anisotropy. Otherwise, only two partial remedies can be recommended: one is to cut the component from the forged bar in such a way that the changes in structure do not coincide with changes of section and do not occur in vital areas; this involves detailed examination of the forged stock. The other is to make the dies by precision casting.

Distortion in *cold treatment* due to thermal effects may lead to warping of awkward shapes. A more likely cause of asymmetrical distortion is the segregation of alloying additions which may lead to an uneven distribution of retained austenite. When the latter transforms the expansion will be greatest in the segregated zones and may cause distortion.

Similar difficulties may be caused in *tempering*. In addition, asymmetry of internal stresses leading to warping may be caused in heating up for tempering as in heating for hardening, by thermal stresses in complicated shapes and by the relief of residual stresses, in this case from the quench.

It would be impossible to attempt to summarize all conceivable causes of distortion and all possible remedies. We have seen that the same effect can cause both shrinkage or growth and that a procedure which will remedy distortion from one cause might greatly accentuate distortion from some other cause. But we must not despair because of this complexity of interactions, we must consider it as a challenge—at least it provides us with a substantial number of handles with which to control and counteract distortion.

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Avoidance of distortion in heat treatment

K. SACHS, Ph.D., M.Sc., A.I.M.

*Three types of distortion are distinguished: Change in volume; change in shape; and warping. Their characteristics, causes and prevention are discussed**

THE EFFECTS of the martensite transformation are superimposed on those of the cooling stresses. The formation of martensite is accompanied by an expansion and obviously takes place at the surface first and in the centre later. The influence of transformation stresses alone are shown in fig. 8*a*. We assume that the block has reached the M_s temperature with no internal stresses or distortion. The martensite transformation and the associated expansion gradually proceed from the outside inwards. The surface layers expand but are restrained by the core which is still contracting; thus the core is in tension and the surface in compression. The austenitic core is much weaker than the martensitic outer zones and is much more likely to yield. This tensile deformation stretches the whole block.

On further cooling, the centre also transforms to martensite and tends to expand; its natural length has been increased by the plastic stretching in the previous stage but it is restrained and kept in compression by the outside layers which themselves are put in tension. The overall effect is further elastic elongation of the block and the stress distribution shown in fig. 8*a* (*bottom*), with the surface in tension and the core in compression.

These stresses and this stretching of the block is superimposed on the distortion produced by cooling stresses which we have already discussed. Some of the more dramatic possibilities are illustrated in fig. 8*b*.

The first is the case of severe quenching leading to upsetting on cooling; it is postulated that the surface is still contracting much faster than the centre. Although the centre has been deformed, it is still in compression and the surface is still in tension. Fig. 8*b* (*middle*) shows that the initial stresses due to transformation tend to cancel residual cooling stresses. Initially, tensile stress in the surface layer is due to its attempt to contract; when it begins to expand, therefore, as a result of martensite formation, the stress is relieved until the partially expanded length of the surface zone corresponds to the natural length of the centre, which is still hot and contracting. Further expansion of the surface will put the core in tension but the resultant stress is likely to be less severe than in the absence of the residual stresses. This is indicated in the figure by the balancing of dimensions, the assumption of zero stress, and the absence of tensile yielding of the core.

Another aspect not shown in the figure must be borne in mind. The temperature gradient is rather severe, leading to considerable time lapse between martensite formation at the surface and in the centre, a factor which allows high stress to build up. This is further accentuated by the fact that a tensile stress favours martensite formation while compression delays it; this is indicated in fig. 8*a* (*middle*) by the M_s temperature, *i.e.* tension raises the M_s, compression lowers it. Obviously, the stress distribution produced by surface contraction helps the surface layers to form martensite earlier and further delays transformation in the centre, so that transformation stresses become more severe.

On cooling to room temperature the core tends to expand, but because it has been plastically upset in cooling its natural length is less than that of the surface and it is in tension, while the surface layers are in compression. Fig. 8*b* (1) indicates that the overall effect is a slight extension, rather less than that to be expected from the transformation stresses. The exact values of the thermal and transformation stresses will determine how near room temperature the 'balance' of dimensions occurs and

*Article based by the author on his lecture given at the Birmingham College of Advanced Technology last January in the series 'Modern developments in the theory and practice of steel heat treatment.' The author is Head of Research Metallurgy Section, G.K.N. Group Research Laboratory, Wolverhampton. The first part of the article appeared last month and is concluded in this issue. Articles based on other lectures in the series will be published in future issues of METAL TREATMENT.

whether the final resultant is a slight upset or a slight extension.

Next, let us consider a very similar case, with a less severe temperature gradient, so that the core is almost as cold as the surface by the time the surface starts to transform. This might correspond to the quenching of a somewhat smaller block. Earlier in the cooling cycle the core has been upset by the contracting outer layers; just before the surface reaches M_s the contraction of the core to its plastically compressed length has overtaken the contraction of the outer zones, which are now in compression, while the core is in tension. The stress system due to transformation is similar in sign, so that the stresses are added together. The surface, which is already being compressed because its natural length is greater than that of the core—it will be remembered that the core has been compressed—now wants to expand as a result of martensite formation, against the continuing contraction of the core. This imposes severe tension on the centre and compression on the outside. The centre is mechanically weaker and yields, this time in tension. This plastic yielding is superimposed on the plastic compression that took place at the beginning of the cooling cycle.

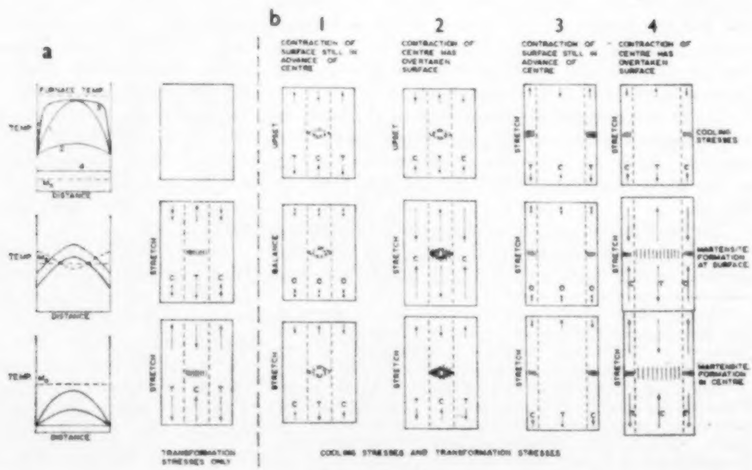
Thus severe elastic stresses are set up, with the core in tension, a complex system of plastic deformation is imposed on the core, but as far as dimensional changes are concerned the stretching due to martensite formation tends to balance the upsetting due to cooling stresses and the resultant distortion depends simply on which of these stress systems is the larger.

On further cooling, when the whole block passes below M_f, the core tends to expand. The extent of the total elongation and the nature of the stress system depends on the natural length of the core relative to the surface, *i.e.* whether the plastic strain was greater in the upsetting or the subsequent stretching stage. In the sketch, fig. 8*b* (2) (*bottom*), the core has elongated slightly and therefore tends to extend the surface zones elastically: the outside is in tension, the centre in compression, and the whole block has elongated a little more.

So far then, the interaction of upsetting due to thermal stresses on heating and cooling and the stretching due to martensite formation may produce either squashing or extension of the block. If the cooling stresses produce a tensile deformation, as might occur in the marquenching of large components, both thermal and transformation stresses combine to extend the block, so obviously there will be a substantial elongation. Two variants are illustrated in fig. 8*b* (3, 4).

The first of these, series 3 in the illustration, corresponds to the early stages of marquenching, when the temperature gradient is very steep. As was shown in fig. 7, rapid and severe shrinkage of the surface zones leads, in these special circumstances, to tensile yielding on the outer layers. The residual elastic stresses are still tensile in the surface and compressive in the centre. These stresses tend to cancel out the stress system induced by martensite formation. The surface layer expands and this expansion relaxes the tension and also, of course, the counter-balancing compression in the core. The dimensions, however, suffer a further longitudinal growth which is added to the stretch produced by the cooling.

As in an earlier example, the stress system has accentuated the time lapse of the martensite trans-



8 *Interaction of cooling and transformation stresses*

formation by raising M_s at the surface. When the martensite formation reaches the core, the latter expands, but its final natural length is still shorter than that of the surface layers, which have extended plastically at an earlier stage; the core still restrains the expansion of the surface and puts it in compression, while the pull of surface layers imposes an elastic tensile stress on the core.

This example has been included because it fits into the sequence logically, although it does not correspond to a particularly probable heat treatment. We have seen that it implies a transformation occurring at a stage in marquenching when the temperature has not yet equalized. Since this defeats the object of marquenching it ought not to happen, but it does not follow that it never occurs. In any case, a temperature distribution of this type may also occur in the water quenching of a large and heavy component.

In normal and sensible marquenching the temperature is equalized just above M_s before the component is air-cooled and martensite is allowed to form. The distortion behaviour in this case is shown in the last sequence of fig. 8*b*. The initial rapid shrinkage has again produced tensile yielding in the outer zones, but by the time the temperature has equalized the core contracts to a shorter natural length than the stretched surface layers. Thus the surface restrains the contraction of the core and puts it in tension, so that the surface itself is in elastic compression.

Martensite formation imposes the same stress system, so that the stresses add up and very high levels are reached. Once martensite has formed in the surface layer, the compressive stresses there cannot cause plastic yielding, and even quench cracking under compressive stress is much less likely than plastic extension of the core. When the core in its turn transforms to martensite and expands, its natural length in relation to that of the surface depends on which zone has been stretched more. The case shown in the figure is that of the outer zone, so that the latter is elongated elastically and the core is in compression.

As already indicated, the possibilities sketched in fig. 8 do not exhaust the permutations and combinations of stress distribution and distortion that can be visualized and are quite possible in practice. Moreover, they completely neglect the influence of incomplete martensite formation on the one hand and retention of austenite on the other. For this reason it is proposed to consider, briefly and far from exhaustively, some of the possible distortions that may occur in practice, how to recognize their source, and how to counteract them.

Some distortions occurring in practice

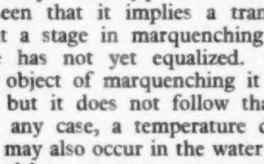
Let us take, first, the case where the upsetting action predominates over the increase in volume to such an extent that the length is shorter after hardening than before, while the lateral dimension has grown considerably. This is illustrated in fig. 9, the original block being shown by dotted lines, the distorted block by full lines. This condition can be recognized by comparing dimensions before and after heat treatment, but it must be emphasized that it is not enough to measure the particular dimension that the design engineer is interested in; dimensions *must* be measured in all three directions. Obvious indications of the predominance of upsetting are (*a*) decrease in length, and (*b*) bulging around the middle.

Possible remedies are outlined in fig. 9*a*. It is assumed that the design dimensions are unalterable; that the selection of the steel would only be altered reluctantly; that any change in quenching medium is only permissible if the mechanical properties are not affected; and that the hardening temperature is the easiest variable to play about with.

To correct the longitudinal shrinkage and bring the bar back to its original length, a number of modifications can be considered. Their object will be either to reduce the upsetting effect, or to increase the amount of martensite formed and thus to balance the upsetting by an increased expansion, or to reduce the amount of retained austenite.

To reduce the upsetting effect, we must attempt to level out the temperature gradient; this is done most effectively by using a milder quench, but a lower hardening temperature may also help to-

a b



a

	Alloy content	*Hardening temperature*	*Quench*
To increase length:			
Reduce upset	Higher	Lower	Milder
Increase martensite	Higher	Higher	More severe
Reduce austenite	Lower	Lower	Chill
To reduce diameter:			
Reduce upset	Higher	Lower	Milder
Reduce martensite	Lower	Lower	Milder
Increase austenite	Higher	Higher	—

b

	Alloy content	*Hardening temperature*	*Quench*
To reduce length:			
Increase upset	Lower	Higher	More severe
Reduce martensite	Lower	Lower	Milder
Increase austenite	Higher	Higher	—
To reduce diameter:			
Reduce upset	Higher	Lower	Milder
Reduce martensite	Lower	Lower	Milder
Increase austenite	Higher	Higher	—

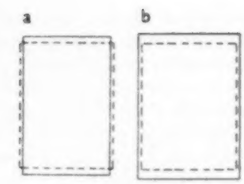
9 *Correcting distortion due to interaction of upsetting and transformation (dotted line before—solid line after heat treatment)*

wards it. The composition of the steel may have some influence in raising the elastic limit of the core during cooling so that it resists plastic deformation. To increase the proportion of martensite in the final transformation product, we can intensify the quench, if that is possible without incurring the danger of cracking; or we can increase the hardenability by raising the austenitizing temperature (and thus the grain size) or the alloy content. We can lower the amount of retained austenite by transforming it to martensite in a cold treatment; or we can reduce the stability of the austenite by lowering the hardening temperature or the alloy content of the steel.

To correct the bulging and bring the diameter of the bar back to its original value, we should have to reduce the upsetting effect as before, but the transformation effects must be reversed because we want to reduce and not increase a dimension; we therefore need less martensite or more retained austenite. The conditions for reducing the upsetting effect are the same as before: a milder quench, a lower hardening temperature, and a higher alloy content to stiffen the austenite. To reduce the proportions of martensite formed we also need a milder quench, or we can lower the hardenability by quenching from a lower temperature or choosing a steel of lower alloy content. The only way to increase the content of retained austenite is to raise its stability by hardening from a higher temperature or increasing the alloy content.

The selection of the actual remedy will depend on practical considerations. The effect of cooling stresses increases with the volume of metal, while the amount of martensite formed decreases. Thus predominance of upsetting over expansion is most likely to occur in fairly large components. If it is desired to correct both longitudinal and transverse dimensions the only possibility is to reduce the upsetting effect. The most effective method would be a milder quench, but since we are dealing with a large block, a milder quench may seriously reduce martensite formation. This will not only make it more difficult to correct the longitudinal contraction, it will also lower the mechanical properties. Thus a milder quench would really only be permissible if a more hardenable steel can be substituted. A promising remedy in this case may be marquenching which might reduce or even reverse the upsetting effect.

If only the length *or* diameter need be corrected, some of the other remedies may be simpler. If the steel contains a fair amount of retained austenite, a cold treatment after quenching may be all that is needed to restore the length. If the bar is not through-hardened, a higher quenching temperature or even a severer quench may be the answer. It is obvious that the best remedy must be chosen



a	*Alloy content*	*Hardening temperature*	*Quench*
To reduce length:			
Restore upset	Lower	Lower	Milder
Reduce martensite	Lower	Lower	Milder
Increase austenite	Higher	Higher	—
To increase diameter:			
Restore upset	Lower	Lower	Milder
Increase martensite	Higher	Higher	More severe
Reduce austenite	Lower	Lower	Chill

b	*Alloy content*	*Hardening temperature*	*Quench*
To reduce length:			
Restore upset	Lower	Lower	Milder
Reduce martensite	Lower	Lower	Milder
Increase austenite	Higher	Higher	—
To reduce diameter:			
Accentuate stretching ..	Higher	Higher	More severe
Reduce martensite	Lower	Lower	Milder
Increase austenite	Higher	Higher	—

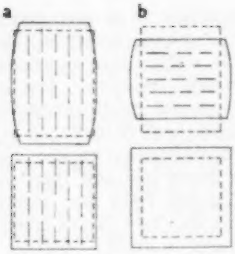
10 *Correcting distortion due to interaction of stretching and transformation (dotted lines before—solid lines after heat treatment)*

by considering the dimensions to be corrected, the limitations of mechanical properties and cost, and the actual structure of the component.

Let us next consider the case where the expansion predominates over upsetting (fig. 9*b*). This is likely to occur in the quenching of smaller components and should be associated with a greater degree of through-hardening than the previous example. Its distinguishing features are (*a*) expansion of all dimensions, (*b*) disproportionate increase in section, and (*c*) slight bulging in the middle. As shown in fig. 9*b*, to correct the increase in length we need to increase the upset or the amount of retained austenite or to decrease the proportion of martensite, *i.e.* the degree of through-hardening. To decrease the section we must reduce the upsetting effect or the amount of martensite and increase the retained austenite.

As in the previous case, some of the recommended measures will help to correct both length and cross-section, but it must be emphasized that we cannot hope to change them both to the same extent. Nevertheless, we may wish to correct one dimension with the minimum distortion of others; in the present case this means reducing martensite or increasing the retained austenite. Both measures would lower the mechanical strength, but an increase in retained austenite might be expected to have less influence on the mechanical properties for equivalent reduction in volume.

If the change to a higher alloy steel is to be avoided, we might try to raise the hardening temperature. This would stabilize the austenite but



a

		Alloy content	Hardening temperature	Quench
To reduce length:				
Increase upset	..	Lower	Higher	More severe
Reduce anisotropy	..	Less carbide	Higher	Milder
Reduce martensite	..	Lower	Lower	Milder
Increase austenite	..	Higher	Higher	—
To reduce cross-section:				
Reduce upset	..	Higher	Lower	Milder
Reduce anisotropy	..	Less carbide	Higher	Milder
Reduce martensite	..	Lower	Lower	Milder
Increase austenite	..	Higher	Higher	—

b

		Alloy content	Hardening temperature	Quench
To increase length:				
Reduce upset	..	Higher	Lower	Milder
Reduce anisotropy	..	Less carbide	Higher	Milder
Increase martensite	..	Higher	Higher	More severe
Reduce austenite	..	Lower	Lower	Chill
To correct cross-section:				
Reduce upset	..	Higher	Lower	Milder
Reduce anisotropy	..	Less carbide	Higher	Milder
Reduce martensite	..	Lower	Lower	Milder
Increase austenite	..	Higher	Higher	—

11 *Correcting distortion due to interaction of upsetting and anisotropy (dotted lines before—solid lines after heat treatment)*

would also increase the upsetting effect a little, which would be helpful in reducing the length, but would counteract the reduction of diameter. The increase in hardenability is not likely to be very detrimental because we have postulated conditions corresponding to a small specimen and a sharp quench, giving virtual through-hardening, so that a further increase in hardenability will not affect the situation.

If it is important to reduce bulging, we might risk a milder quench to reduce the upset and take this a little further by lowering the hardening temperature. If the mechanical properties deteriorate too badly, we can use a steel of higher alloy content which will restore the martensite, but will further reduce the upsetting effect and favour the retention of austenite.

For our third example we chose the marquenching of a heavy block under conditions where an extremely severe temperature gradient reverses the normal upsetting effect of the cooling stresses and stretches the block instead (fig. 10*a*). The expansion due to martensite formation is added to this, and the gross effect is elongation and lateral shrinkage of the bar, which should make this more easily recognizable. Both can be partially corrected by restoring the upset. This requires some reduction in the severity of the temperature gradient in the surface layers, a milder quench, or a lower hardening temperature, perhaps combined with a lower alloy content to give a weaker core, more likely to yield in compression. If lowering the hardening temperature is ineffectual, a milder quench can be achieved by substituting a salt bath for a lead bath or even trying an interrupted oil quench.

Another expedient is the substitution of air cooling for marquenching, compensating for the loss of mechanical strength by using a steel of higher hardenability. Separate adjustment of the major dimensions can be achieved by modifications of the martensite and austenite contents in ways with which we are by now familiar.

In the next case (fig. 10*b*) the temperature gradient in marquenching produces slight stretching, but the expansion on martensite formation is sufficient to increase the section as well as the length. The condition can be recognized by a disproportionate increase in length and the absence of bulging. Both length and cross-section can be reduced, but not to the same extent, by an increase in the amount of retained austenite or a reduction in through-hardening; the length can be reduced at the risk of further growth of the section by restoring the upsetting effect, while the diameter can be reduced at the expense of further elongation by accentuating the stretching.

These conditions of distortion correspond to the marquenching of a fairly small block and might be cured by oil quenching to avoid the stretching effect in cooling and the use of a higher hardening temperature to favour retained austenite.

The last two cases we shall consider involve structural anisotropy of dimensional changes. The carbide stringers present in most tool and die steels tend to favour extension of the block parallel to the stringers and this is compensated by contraction at right angles.[3]

Apart from the presence of carbide stringers, the block of steel is still subject to the usual thermal stresses and the overall volume change associated with the formation of martensite and I shall now deal very briefly with their interaction. The discussion will be confined to the interaction between anisotropic dilatations and thermal upsetting, and will not include the much less frequent case of stretching due to cooling stresses.

If the carbide stringers are parallel to the length of the bar, the anisotropy counteracts the shortening due to upsetting and the bar is lengthened (fig. 11*a*). The cross-section, originally square, has grown due to the upsetting effect, but the alignment of carbides in one direction has led to increased expansion in this direction and less expansion at right angles so that the section has become oblong. All dimensions can be partially

corrected by any steps taken to reduce the anisotropy: changes in composition to reduce the amount of carbide in the structure, a higher austenitizing temperature to take more carbide into solution, and slower cooling which diminishes the effect of the carbide stringers.

Other ways to reduce both the length and cross-section are to increase the amount of retained austenite or to reduce the amount of martensite. However, if the properties are to be maintained, reduction in martensite is hardly permissible, particularly if any compensation by increasing the carbide content would accentuate distortion. Clearly the first step to try is to raise the hardening temperature which should favour retention of austenite by dissolving alloy carbides and thus diminishing the anisotropy. If the length is not terribly important a milder quench would help greatly in correcting the cross-section, but this may cause trouble with mechanical properties.

If the carbides run transversely (fig. 11*b*), the upsetting will be greatly accentuated, the length of the bar will decrease considerably, and the cross-section will greatly increase but will remain square. All dimensions can be partially corrected by steps to reduce both anisotropy and the upsetting effect, but the selection of the actual remedy depends on which dimension we want most to correct.

If we want to reduce the square section, we must reduce the anisotopy by using a milder quench and a higher hardening temperature: only experiment can show whether the higher temperature raises hardenability sufficiently to compensate for the milder quench and leaves the amount of martensite unaffected. If it does, its effect on austenite retention will help to shrink the cross-section but will also accentuate the longitudinal shrinkage.

If the length is the most important dimension we should try a milder quench to reduce upset and anisotropy and recover the loss of properties by transforming some retained austenite in a cold treatment. But if the length is the most important dimension it was foolish to machine the component with the carbides running transversely in the first place. Where anisotropy plays an important part in distortion, intelligent selection of the raw steel for machining can help a lot in reducing distortion.

Distortion in quenching has been considered at some length because this is the stage in the heat-treatment sequence where distortion occurs most frequently and where a large number of influences combine to cause distortion.

Symmetrical distortion in cold treatment and tempering

Cold treatment Transforming retained austenite to martensite improves the mechanical properties and causes an expansion which may correct shrinkage in quenching. In a large component that is not hardened through, cooling to −78°C. may cause upsetting: the surface layers shrink first and will tend to compress the core. The core is weaker than the surface and yields in compression, so that the whole component is upset, as in rapid cooling from the hardening temperature. The temperature difference is much less and there is no need for high heat transfer rates and severe temperature gradients, so that effect is not likely to be severe and is fairly easy to avoid.

The expansion accompanying martensite formation will also cause transformation stresses, but will counteract the purely thermal contraction in cold treatment. Incomplete through-hardening implies that only the surface layers contain retained austenite so that when the cold treatment is complete and the component is back at room temperature the surface will have expanded more than the core.

Many of the factors causing distortion can operate during *tempering*. We have shown that hardening leaves complex systems of internal stresses in the component and these will be relieved by relaxation or yielding on heating for tempering. The tempering of martensite leads to contraction, the tempering of lower bainite to rather less contraction, the precipitation of carbides from retained austenite may cause expansion. Since these phases will not be distributed uniformly and the reactions do not occur simultaneously, there are many opportunities for internal stresses to be set up.

Cooling after tempering may lead to upsetting, particularly if oil quenching is used to avoid temper brittleness. Austenite impoverished in alloy and carbon content by the precipitation of carbide during tempering will be less stable and will transform to martensite on cooling—the expansion may cause distortion. The decomposition of austenite on tempering can be used for correcting distortion in the same way as cold treatment, but may require a higher tempering temperature than the mechanical properties will stand. Selected dimensions can be controlled by sequential tempering until the desired size is reached.[4]

Warping

The most obvious and dramatic form of distortion is warping. It is also most difficult to overcome by grinding and the one that causes most trouble generally. An example has already been given of warping caused by variations in cooling rate of large and small sections.

In general, it can be said that warping is produced by causes similar to those responsible for symmetrical distortion and there is no need to reiterate them. Warping results from the uneven distribution of the internal stresses induced by

these various mechanisms. We may now run through the various heat-treatment operations and consider briefly what causes warping and how it may be avoided.

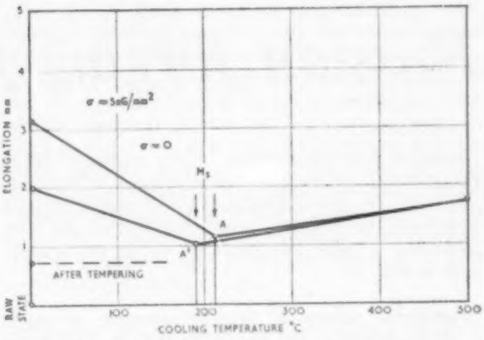
We have seen that in *heating up for hardening* the relief of internal stresses may cause distortion. It is very easy for a machining operation to remove metal carrying a residual stress and thus produce a completely asymmetrical redistribution of internal stresses. If a component with such a stress system is heated too rapidly the distortion produced by local yielding under these stresses can cause severe warping.

Heating also gives rise to thermal and transformation stresses and these can be asymmetrical if the shape is irregular. Thus the pulley we saw in fig. 4 could buckle during heating because the web would get hot quicker than the other parts and would tend to expand against the restraint of the boss and rim. As in the case of symmetrical distortion the remedy is slow and even heating.

Warping during heating up may also occur due to sagging, if the component is badly designed or inadequately supported in the furnace, or due to an uneven temperature distribution in the furnace. It is surprising how often practical problems can be solved by a little attention to elementary good practice. In extenuation, it might be said that it is by no means easy to build a furnace of reasonable size with a uniform temperature distribution throughout the useful space. But life is not made any easier by people who open furnace doors to light cigarettes.

In *quenching*, again, all forms of distortion can turn to warping if the stresses produced are not distributed symmetrically. This is particularly likely to happen to components of complex shape having parts of different sections. Warping may also be caused by bad quenching practice. Long thin components must be quenched vertically. Heavy components must be moved about in an oil quench to prevent the formation of stable bubble films. Conversely, something can be done to counteract warping by special quenching techniques, jigs, etc. The pulley in fig. 4 might be less susceptible to warping if, instead of immersing the whole thing in the quench tank, we spin it round fairly fast with only the rim in the quenching medium. This would bring the cooling rates of the rim and web closer together.

Experienced heat treaters, even those who have never heard of T-T-T curves, often use an interrupted quench to allow the temperature to equalize before the martensite transformation starts. The same thing done under more controlled conditions and with the T-T-T curve as a guide is, of course, marquenching; this practice greatly reduces the



12 *Effect of bending stress on the expansion associated with martensite formation*

incidence and severity of warping in the quenching of irregular shapes.

Long thin components may easily bend due to sagging, or slight asymmetry of the structure or cooling conditions. Unless they have to be used in a very hard and brittle condition they may sometimes be salvaged by cold straightening after tempering. An interesting method has recently been proposed in a Polish patent for the straightening of bent broaches of high-speed steel during air hardening.[5] For air cooling they are placed in a jig. Just before the M_s temperature is reached, a light load is applied in a direction tending to straighten the broach, but only sufficient in magnitude to impose elastic bending stresses.

In fig. 12 dimensional changes are shown plotted against the temperature, with and without a slight stress. The stress raises the M_s temperature on the tension side so that the concave part which is under tension starts to expand and the martensite transformation does the straightening rather than the externally applied load. The author claims that the part transformed under tension reaches a greater length at room temperature and attributes this to orientation of the martensite needles and therefore anisotropic expansion under the influence of the tensile stress.

Another cause of warping in quenching is irregularity in compositional or structural variations. In a carburized component the difference in carbon content between surface and core will give rise to differences in expansion behaviour and transformation characteristics and this may well lead to internal stresses. If the component or the carburized layer is irregular in shape, these stresses will be asymmetric and will cause distortion. The same thing can happen as a result of accidental decarburization (this is more likely to be irregular

continued on page 316

Russian forging journal

Abstracts from the Russian forging journal—' Kuznechno - Shtampovochnoe Proizvodstvo,' January, 1961, **3.** *This is the third year of this journal devoted specifically to forging. Short abstracts of the more important articles are given in* METAL TREATMENT *each month.*

Hydrostatic testing of biaxial extension of pipes. L. A. RUBENKOVA. Pp. 3–4.
The article outlines the theoretical and mathematical bases of such testing, and indicates that it may suitably be used for testing metal intended for the production of pipes for gas and other pressure mains.

Drawing of double-walled components. M. N. GORBUNOV and TAN YUN-SI. Pp. 5–8.
Problems of such drawing and calculation of the limiting parameters are discussed.

Viscosity of lubricants and the efficiency of their action during the drawing of thin sheet steel. V. I. KONOPLINA. Pp. 8–9.
The investigations conducted show that with the increase in the viscosity of gun oil through the addition of talc there is a decrease in the drawing force during the drawing of cylindrical components and consequently an increase in the efficiency of operation of the lubricant.

Automatic production lines for bolts and screws. A. N. GLADKIKH, N. I. MASLENNIKOV and P. P. FARAFONOV. Pp. 9–11.
Two lines at the ' Red Etna ' works in Gor'kovsk are described.

Bending with underpressure. P. G. SHISHLAKOV. Pp. 12–13.

Forging of uranium (review based on literature data). P. I. SEREDIN. Pp. 14–18.

Automatic forging rolls. A. F. BALIN and P. KH. VALITOV. Pp. 18–23.
A new improved design of automatically operating forging rolls is described. Three stands, each of which has a pair of single-pass rolls, are mounted on a common baseplate, and the billet is fed into the work rolls by a pusher mechanism. In a continuous production line it is producing billets for forging automobile engine connecting rods in 4–5 sec. in three passes. A full description is given of the complete unit. The rolling rate is so high that the loss in temperature of the billet is negligible. To compensate the wear on the work rolls and for roll setting, eccentric bushes are used for reliable setting of the distances between the centres of the leading rolls.

Calculation of hydro-pneumatic overload switches of sheet stamping mechanical presses. V. V. MAGAZINER, M. M. ROZENBLAT and V. I. SOKOV. Pp. 23–28.

Simplification of the calculation of the movement of the cross-head of a hydraulic press with accumulator. S. N. BELYAEV. Pp. 29–30.

Choice of a crankdrive press for drawing. A. G. OVCHINNIKOV. Pp. 30–33.

Determination of the elements of the triple-lever articulated mechanism of double-action presses. N. P. KATKOV. Pp. 33–36.

Methods of improving the operation of compartment heating furnaces. A. U. PUGOVKIN. Pp. 36–39.
After outlining these methods, model work is described which has led to the construction of a new vertical, recirculation furnace which has given good results under production conditions. Heating was very even and fuel consumption greatly reduced.

Bottom ejection systems for extrusion forging. YU. P. VOLIK and V. V. BOITSOV. Pp. 40–43.

A universal jig for group stamping of cylindrical components in machinery construction. E. I. LEBEDEVA. Pp. 43–46.

Stamping of Duralumin pistons. V. G. MAKAROV. Pp. 47–48.
Old and improved methods of stamping these pistons are compared. The new flashless method reduces the weight of the stamping, decreases variation in wall thickness to 0·5 mm. and increases output, one operation being eliminated.

February, 1961

Effect of friction of the magnitude of the penetration force of a rounded ram. A. D. TOMLENOV. Pp. 1–3.

The problem of the determination of calculated forces of mechanical hot-forging presses. I. I. GIRSH. Pp. 3–6.
An attempt to establish a relationship between calculated press force and the diameter and height of a comparison forging is outlined on the basis of experimental data.

Mechanization and automation of spinning lathe operations. V. G. KONONENKO and YU. A. BOBORYKIN. Pp. 6–8.

An investigation of spread during the cogging of forgings. I. M. BALYASNYI. Pp. 8–15.
Practical verification of the results of the investigation of the spread of forgings and of the suggested recommendation for its calculation was carried out during the forging of tyres on a 10,000-ton press. The tyres were forged in accordance with the new technique combining three operations for one heating of the billets—upsetting, piercing and cogging to forging dimensions. Intermediate upsetting and heating were cut out. The new technique has been adopted as standard production practice.

Effect of the shape of the tools and of the amount of compression on the distribution of longitudinal deformations over a cross-section of the forging. G. M. BELKOV. Pp. 15–20.
Multi-layer lead model ingots were forged between plane and angled tools and combinations of these. Surface and internal deformations, the homogeneity of their distribution, and their mutual relationships were studied.

The problem of the productivity of press equipment for plastics. A. I. ZIMIN and A. S. EZZHEV. Pp. 20–23.

Calculation of the stresses occurring in the walls of crossheads of presses close to the point of fixing of the columns. S. I. BLINNIK. Pp. 23–25.
The calculation method outlined enables the work done by a crosshead close to the bushes to be calculated and the required thickness of its walls to avoid the occurrence of cracks.

The use of friction switch clutches as safety devices. E. N. IZOTOV, M. M. ROZENBLAT and V. I. SOKOV. Pp. 25–28.
Consideration is given to two opinions on friction switch clutches in conjunction with friction safety devices and without them. It is found that safety devices placed in the flywheel of a press do not ensure protection of all parts of the press against heavy overloading in the torque of the press. The efficiency of such devices is dependent on their suitable location, *e.g.* in some instances on the crankshaft. In addition there must be constancy of the air pressure in the pneumatic cylinder of the press under various working parameters, the specific pressure on the friction surfaces must be such that the dynamic friction coefficient and the static friction coefficient are equated, and the moments of inertia of the main drive components must be at a minimum. Relationships were established between the static coefficients of friction and the specific pressure for friction pairings of cast iron-fibre and steel-Ferodo.

A method of calculation of the electrical drive of crankdrive presses. V. P. MECHANIK. Pp. 29–33.
Existing methods of calculation are based on numerous assumptions which often lead to substantial errors in calculations. Consideration is given to a method which avoids some of the generally accepted assumptions. Use is made of a graph of the movements of the forces of resistance on the crank shaft, constructed with allowance for the mechanical losses in the intermediate transmissions of the press. Amongst the initial data used are also the work done in deformation by the press and the work done in engaging the clutch.

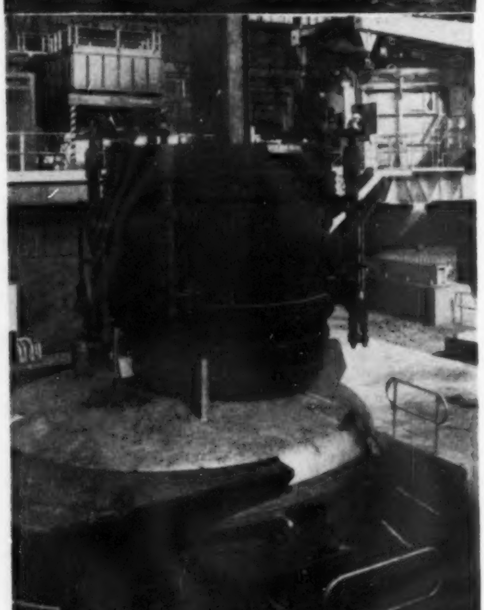
An automatic gas furnace for non-oxidizing heating before stamping. V. V. BOGATYREV. Pp. 34–37.
The new furnace is designed as one unit of an automatic stamping line for stamping automobile body components. The furnace is fired with natural gas at 1,100°C. for heating billets to 900°C. in 7–8 min. (in the first zone to 500–550°C. and in the second to 900°C.). Air is preheated in recuperators to 500°C. and combustion is at an excess air coefficient, $d = 0{\cdot}45$. There are several new features in the design.

A technical and economic analysis of various methods of heating forging billets. V. S. BYALKOVSKAYA. Pp. 37–41.
Precepts are outlined for the choice of heating methods.

A die with an articulated ram and a spring delivery system. M. M. KUZNETSOV. Pp. 42–43.

A special press for stamping sheet. YU. A. SKUCHILIN and V. F. FEDORKEVICH. Pp. 43–44.
A 2,500-ton, double-crank, single-action press for cold stamping of sheet components of large dimensions and appreciable length is described.

An automatic delivery mechanism for coiled wire or strip into a cutting die. S. I. AZAROV. Pp. 45–47.



TOP *Topping the 80-ton furnace*

BELOW *Ladle in position over vacuum-casting tank*

Vacuum-casting installations at River Don Works, Sheffield

THE largest electric arc-melting furnace in the U.K. was commissioned last month at the River Don Works of English Steel Corporation Ltd. for the production of special alloy steels. This furnace, which has a shell diameter of 21 ft. and an electrical rating of 20 MVA., is capable of melting an 80-ton charge in less than 3 h.

The furnace has been designed and built by Birlec-Efco (Melting) Ltd. for the requirements of English Steel Corporation Ltd. Power is supplied to the furnace by a 20-MVA. transformer of A.E.I. manufacture, which has a resistance transfer on-load tap change mechanism especially suitable for arc-furnace duty. A 10% auxiliary reactor arranged for on-load switching is also provided to limit power surges during the initial melting-down period. The main switch employed is an oil circuit breaker, having a 750-MVA. rupturing capacity, and is manufactured by A. Reyrolle & Co. Ltd. Specially designed water-cooled low-voltage flexible cables are used to carry current from the transformer to the furnace. During the melting period this current is approximately 28,200 amp. per phase.

Each electrode arm is counterbalanced by means of four pneumatic cylinders, enabling relatively small motors to achieve rapid accelerations and high speeds with a minimum of inertia. These motors are of special design to operate with a well-proved A.E.I. amplidyne electrode controller. The electrode diameter is 20 in.

The furnace shell incorporates a new feature in that it is split horizontally in two halves just above the slag line. The upper half is detachable to enable the side wall brickwork to be replaced in the shortest possible time. To accomplish this a spare upper half, which has been previously bricked, can be lowered into position in a similar manner to the way an arc-furnace roof is replaced in conventional practice.

The furnace has been designed for the production of special alloy steels for casting in vacuum. The vacuum-casting unit is of Bochumer Verein design. By combining charges of other furnaces with the charge of the 80-ton furnace, it will be possible to vacuum cast ingots up to 180 tons in weight. Ingots so produced are extremely low in hydrogen content, and forgings made from these ingots are particularly suitable for highly stressed forgings for the electrical, chemical and general engineering industries and for other high-duty applications.

Applications of electron bombardment heating in metallurgy

*Electron bombardment heating has found many applications in research and industry. The technique permits intense and controlled heating over small areas without contamination of the material being heated. The application of this form of heating to such processes as zone melting, single crystal preparation, welding and ingot production is described**

THE HEATING EFFECT produced by electron bombardment has been known ever since the first use of electron devices. It was used for heating laboratory specimens over 25 years ago. In the last ten years electron bombardment heating has been applied to many problems, in some cases on a large scale (>100 kW.), and is now a rapidly developing technique.

Electron bombardment heating requires the production of a controlled stream of electrons between an emitter and the material to be heated. Obviously this is only feasible *in vacuo* as at higher pressures the mean free path of electrons is too low to be useful. In practice the process is feasible at pressures below about 10^{-3} mm. of mercury.

Floating-zone melting

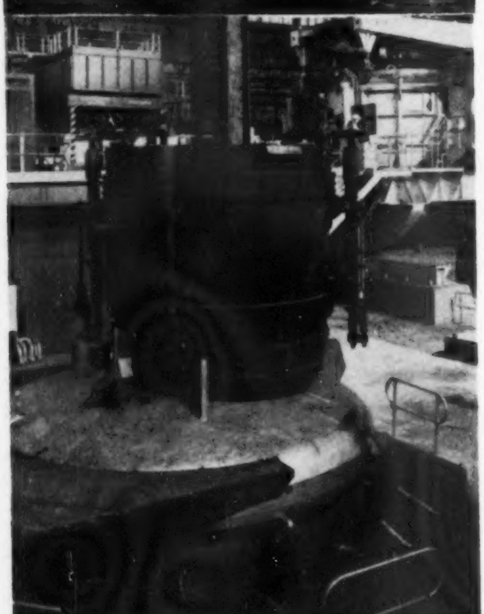
Electron bombardment heating is at present widely used for zone melting on a laboratory scale. It is most important in the zone melting of reactive and refractory metals, which cannot be melted in crucibles without contamination so that the technique of floating-zone melting is used. In this technique a narrow region of a vertical rod is melted and the molten region is held in position by the surface tension forces. The zone can then be moved along the rod either by passing the heater along the rod or by passing the rod through the heater. As the zone is held in position by surface tension forces, the length of the zone is limited (less than about 0·5 in. is a typical figure) and is related to the thickness of the sample. Small zones such as these can be easily melted with a simple electron-bombardment heater.

In such a furnace used at Aldermaston, a rod specimen is rigidly supported on a frame in a vacuum chamber. The small electron-bombardment heater surrounds a short length of the rod and is co-axial with it. Typical heaters are very simple, consisting of a single-turn filament heated by a low-voltage supply. The filament is held at a large negative potential (≈ 1 to 10 kV.) relative to the rod; in alternative designs either the rod or the filament may be earthed. Two metal plates on either side of the filament constitute a modulator and are held at the same potential as the filament which is commonly made of tungsten wire.

The vacuum systems are conventional, consisting of a rotary pump backing a diffusion pump and vapour trap. The pumping speed of the system needs to be at least 20 l./sec. and a speed of up to about 100 l./sec. or more is desirable if very gassy specimens are to be treated.

Specimens to be zone-melted in this type of furnace must obviously be in the form of rods and sometimes must be formed into this shape. Two alternative techniques of mechanical forming are currently used: arc melting and powder pressing. These are useful when the metal evolves considerable quantities of gas on melting. In the electron bombardment furnace the gas bubbles out of the metal with sufficient violence to blow off some of the material. Unless very thin layers of the specimen are melted in successive zone-passes the loss of metal may be considerable. Most of this gas can usually be removed by argon-arc melting the material beforehand; alternatively, if the metal is available as powder, it can be pressed into suitable rods and these can then be heated gradually in the zone-melting furnace so that most of the gas is released whilst the pores in the compact are interconnected.

*Abstract of a paper by N. F. Eaton, B.SC., PH.D., A.I.M., Research Department, Associated Electrical Industries (Manchester) Ltd., D. B. Gasson and F. O. Jones, B.SC., AEI Research Laboratory, Aldermaston, presented at a symposium on 'User experience of large-scale industrial vacuum plant,' arranged by the Institution of Mechanical Engineers. The complete article was reproduced in 'AEI Engineering,' July/August, 1961, by whose courtesy the present extracts are given.



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Zone melting is used in the laboratory for refining, homogenizing of alloys and growing of single crystals. To date, the electron bombardment furnace has been used principally for growing single crystals of metals. Single crystals of an extensive range of metals have now been produced, including even some of those with a phase change such as titanium and cobalt. Homogeneous single crystals of at least one alloy (molybdenum/50% rhenium) have been produced. In many cases the single crystals so produced are much purer than the starting material because of the volatilization of impurities with a higher vapour pressure than that of the parent metal. For example, niobium crystals have been produced with a hardness in the region of 40 to 50 V.H.N. from material with a hardness of 150 V.H.N. or greater, and tantalum crystals with a hardness of about 50 V.H.N. from tantalum powder which arc melts to a metal with a hardness of 500 V.H.N. In both these cases the hardness is an indication of the oxygen, carbon and nitrogen contents of the metal. Tungsten crystals produced from tungsten powder have appreciable ductility and can be bent at room temperature.

The technique is not used extensively for zone refining as in many cases the lowest speed with which the zone can be moved is limited to a relatively high value, in the region of 3 in./h. Zone refining often needs speeds considerably less than this and the vapour loss from many metals is prohibitive at these very slow speeds because of the long time the metal spends at temperatures *in vacuo*. The purification due to vacuum melting, especially with the high-melting-point metals, is, however, so great that this limitation is not necessarily a great disadvantage.

This vapour loss is also important in the growing of single crystals of alloys, as in the homogenization of alloys, where the more volatile constituent will be lost preferentially. This limits the alloys treatable to those in which the vapour pressure of all constituents is less than about 10^{-3} mm. at the melting point of the alloy.

An unexpected defect found in some single crystals is microporosity. In some cases gas bubbles are formed which adhere to the solidifying surface and are entrapped as it moves forward. Such porosity can be avoided if the metal involved is melted in an argon-arc furnace before zone melting. Sometimes even this is unnecessary if the zone is moved upward so that the gas bubbles float away from the solidifying surface. These pores contain gas at an appreciable pressure (up to about $\frac{1}{2}$ atm. or more) because of the surface tension forces at the surface of the small pore. They would not occur if the metal was in equilibrium with the vacuum, as the partial pressure of the gas over the metal would then be very low and the pores would be unable to nucleate. This suggests that to take full advantage of the vacuum melting process all parts of the metal should be exposed to the vacuum at some time, that is, some form of stirring or agitation of the melt may be necessary.

Despite these relatively minor defects, the technique has been proved a valuable tool in the laboratory and single crystals produced in this fashion are widely used for research. In the future its use may be extended on to a larger scale by using slab specimens in which a long horizontal zone is melted by a beam of electrons focused to a line.

A Czochralski crystal pulling furnace

The floating-zone technique has been used to grow crystals of semi-conductor materials, but for some purposes the Czochralski technique is more suitable, for example if the starting material is in the form of lumps or if there is only a limited amount of material available. A special furnace using electron-bombardment heating has been built to grow silicon crystals using this technique.

Silicon is a material which can easily be contaminated by trace impurities originating from materials used as heaters and crucibles, and an over-riding consideration in the design of apparatus for growing single crystals is the need to reduce the effects of contamination of the material to an acceptable minimum. Sufficient experience had been gained with a pulling furnace for preparing silicon crystals, using eddy-current heating, to consider what advantages might be gained by substituting electron bombardment as the heat source. This entailed forming a molten pool resting on and entirely contained by an underlying layer of parent solid material, instead of the more usual method of containing the material in a refractory oxide crucible. Since the normally available high-purity raw material is in randomly shaped lumps, this favoured the use of specifically directed electron beams, rather than the arrangement in which the cathode is positioned close to the material being heated.

The total power required in this furnace was based on that required to operate a conventional pulling furnace which used 0·5 Mc./s. eddy-current heating. After allowing for various losses, it was considered that 2 kW. of beam power would be necessary and for larger-scale operations 5 kW. of beam power should be available.

Actual crystal preparation is carried out by converging each beam on to the middle of a piece of silicon weighing about 20 g. which rests on the water-cooled hearth. As the molten pool starts to form, the electron beams are deflected outward either individually or collectively to enlarge the

pool. The beam powers are then adjusted by moving the filament so that a stable molten pool is formed, and the single crystal seed is dipped into the middle of the pool. When the seed tip is wetted by the melt it is withdrawn at a speed of approximately 2 mm./min.

Pulling can be continued until the edge of the molten pool approaches the base of the charge, at which stage the pulled crystal is raised clear of the melt and the beams are deflected inward to impinge on the crystal to reduce heat loss consequent upon breaking contact with the melt.

The significant property of crystals prepared by this 'crucible-less' pulling technique *in vacuo* is that the oxygen content is greatly reduced. Optical transmission measurements made at 9μ wavelength give a measure of the oxygen concentration, and the reduced oxygen content in these crystals gives a negligible absorption peak compared with a measurement on a crystal prepared by more conventional methods.

Development of this type of multigun assembly may find application in the metallurgical as well as the semiconductor field. Thus in floating-zone recrystallization of bar stock a modification of this layout may be of a form in which the bar to be recrystallized is held stationary while the relatively long-range electron beams can be deflected in unison to pass a molten zone along the bar. Cross contamination of the cathode and material being recrystallized is reduced to a minimum by this arrangement.

Electron-beam melting

Both high-frequency (h.f.) induction and consumable electrode-arc furnaces are now established processes for the vacuum melting and casting of metals. However, for high-purity melting, h.f. induction has limited application due to the problem of contamination from mould and crucible materials. The development of consumable-electrode arc furnaces with water-cooled copper moulds was a big step forward in minimizing impurity pick-up. However, in a consumable-arc furnace the normal method is to melt the ingot into a fixed mould, starting at the bottom and feeding in the electrode. Satisfactory arc conditions require fairly short arc lengths, and thus the release of volatile impurities is restricted by the confined geometry of the system. Further, melting is necessarily continuous, as the heat input depends on the maintenance of the arc which consequently means continuous feed of electrode. Prolonged exposure of the molten surface to facilitate release of volatile impurities is thus not possible. Further, the addition of alloy material or reclamation of scrap by hopper feeding is unsatisfactory owing to possible shorting of the arc. All material to be melted must therefore be fabricated to electrodes.

Electron-beam furnaces can overcome, to a large extent, the limitations of consumable-electrode arc furnaces. By suitable electron optics the width, depth and temperature of the pool can be controlled by power or focus adjustments to a greater degree than is possible with an arc. Since this can be attained irrespective of the rate of addition of material to be melted, the method ensures optimum condition for release of volatile impurities, less likelihood of re-entrapment of dross, minimum trouble from porosity and improved control of solidification compared with the arc-melting method. Further, in addition to the melting of solid consumable electrodes, the production of vacuum-melted ingots is possible from swarf or small scrap material without the need for electrode fabrication. The vacuum reclamation of scrap material directly into ingots is particularly promising.

A simple electron-bombardment melting furnace has proved valuable in the laboratory and has been used to purify samples of electrolytic uranium. In this metal the principal impurity is uranium oxide, which is insoluble in the uranium and considerably less dense. As the melt is not stirred when melted by electron-bombardment heating, the oxide floats to the surface, leaving a much cleaner sample.

Electron-beam welding

The use of a focused beam of electrons for welding applications provided the first convenient means of welding in a vacuum. Inert-gas tungsten-electrode arcs become unstable below 100 mm. pressure and, while magnetic means have been used on a laboratory scale to give stable arcs at pressures down to 10^{-2} mm. of mercury, this has not been developed for welding applications.

As will already be appreciated, vacuum welding is of particular importance for materials such as tantalum, niobium and molybdenum, since the properties of these materials are adversely affected by impurity pick-up in conventional inert-gas welding. While welding in inert atmosphere chambers has solved the weld embrittlement problem in tantalum and niobium, in the case of molybdenum sufficient impurities to cause embrittlement are picked up from inert atmospheres of the highest purity available. Butt joints in molybdenum sheet, electron-beam welded, however, have been bent cold. This has never been possible with inert-gas welds. The reason for this success with molybdenum is twofold. Firstly, impurity pick-up from the welding atmosphere is minimized. Secondly, the melting of the molybdenum in a vacuum allows release of volatile molybdenum oxide, thus purifying the weld bead with respect to oxygen. Oxygen has been shown to be the element

most responsible for the embrittlement of molybdenum, and its concentration in molybdenum has to be reduced below a few parts per million to lower the ductile to brittle transformation below room temperature.

The high degree of control of the energy input to the weld by control of the electron beam makes the welding process particularly attractive. The degree of control of welding variables is well illustrated by the excellent appearance of the welds obtained by this method. Variation of the beam current, voltage or focus allows a high degree of heat-input control during the actual welding operation. Low heat-inputs over a large area of the work, by the use of a diffuse focus, enable pre- and post-heating of the weld area *in situ* to an extent that is not possible in arc welding. This is of particular use in materials which are susceptible to hot cracking or which benefit from heat treatment after welding. For example, no difficulty was experienced in welding tube-end-cap joints in beryllium by use of this technique.

Application of the facility of defocusing the electron beam has provided an interesting method of sealing fuel element cans for nuclear reactors. Workers at Harwell have developed a technique in which the end-cap is first welded with a fine focused beam which gives only localised heating. The beam is then defocused to give general heating over the specimen. This melts a pre-placed ring of braze below the weld, forming an additional seal.

A further significant advantage of the electron-beam welding process is that very high heat inputs per unit area are possible by fine focus of the electron beam. By suitable electron optics a focused spot as small as 0·005 in. dia. may be obtained. The practical result of this is that greater penetration is possible for a given pool area. Penetration to width ratios, approaching 10 : 1, have been achieved in butt welding, a value far greater than is possible with an arc. The criterion here is to use high kV. and low beam currents. The electron optics resemble those in electron microscopes. German and American equipments of up to 150 kV. are under development.

The attainment of high-power fine-focus heat inputs facilitates the welding of very thin material. Conversely, in the welding of thick sections, high penetration to width ratios result in a much smaller volume of coarse-grained weld material for a given penetration, and an associated narrower heat-affected zone. These two factors have resulted in improved mechanical properties in zirconium alloys, where the impact strength of electron-beam-welded material is twice that of argon tungsten arc welds. A comparison with unwelded parent metal showed the improvement due to lack of atmosphere contamination was of less importance than the reduced weld area and heat-affected zone.

A very wide range of metals has been welded by this method. The main difficulty encountered has been concerned with vapour release into the chamber due to either the high vapour pressure of the metal being welded, or to the constituents it may contain which may be alloy additions or impurities. If the pressure rises in the chamber, electrical breakdown by discharge can occur. This is minimized by the use of a pulsed heat-input, together with a vacuum pumping system of high capacity which allows the metal atoms to be pumped out of the chamber during the 'off' cycle.

Electron bombardment with both focused and diffuse beams is also used as a method of heating for vacuum brazing. In brazing, as in welding, the advantages provided are those of a localized, easily controlled heat source.

It is in the field of atomic energy that electron-beam welding has found most application and consequently the field in which most advantages have been demonstrated. It is used on a commercial scale by the French and the Americans for the welding of the fuel elements of some of their reactors. Mechanization to manipulate a multi-charge of fuel elements within one vacuum chamber is used. Indeed, it has been suggested that from an economic standpoint the electron-beam welding process has significant advantages over the vacuum-purged inert-gas chamber arc welding normally used for special materials or applications. In terms of consumable items such as electrical power, inert gas used, etc., the electron beam process is some 35 times cheaper.

Obviously the main limitation to the process is the necessity to contain the component to be welded within a vacuum chamber. However, this problem is not unsurmountable and so the suitability of the process for a particular application has to be decided on the basis of the economics, the welding conditions necessary and the properties of the weld made by the electron-beam process compared with those of the welds made by other methods.

Ipsen Industries Ltd.

Ipsen Industries Inc., of Rockford, Illinois, in association with Ipsen Industries International G.m.b.H., of Kleve, West Germany, manufacturers of atmosphere and vacuum automatic heat-treating equipment and associated plant, have formed a British company, Ipsen Industries Ltd., 53 Victoria Road, Surbiton, Surrey, to facilitate further expansion of sales and service in Great Britain. Mr. L. E. Plimley, A.M.I.E.E., has joined the British staff and operates from 38 Monckton Road, Quinton, Birmingham 32 (Woodgate 4941). Several Ipsen installations are already in operation in this country and many more are in hand for the future.

New engineering laboratory at BWRA

Last month saw the opening of the new engineering laboratory of the British Welding Research Association at Abington Hall, Cambridge, by the Rt. Hon. Lord Mills, P.C., K.B.E. An exhibition of research in progress also took place on this occasion and the following account gives brief details of some selected investigations

THE NEW ENGINEERING LABORATORY of the British Welding Research Association at Abington is now completed and is fully occupied with the exception of one wing. It is a two-storey building providing an area of 15,000 sq. ft. on the ground floor and 8,200 sq. ft. on the upper floor. Like the other post-war buildings at BWRA it is a plastically designed all-welded steel-frame structure. Plastic design, which has ultimate load as its design criterion, has been developed in this country for rigid-jointed frameworks of mild steel which carry load by virture of the resistance of their members to bending action. The accuracy of the plastic design methods, unlike that of the elastic or 'classical' methods, is not affected by the occurrence of residual stress, sinking of supports, stress concentrations, etc. The principal advantage of the plastic method is the saving in steel which it allows. The building was designed by H. C. Hughes and P. Bicknell, of Cambridge, and constructed by Rattee & Kett Ltd., also of Cambridge.

The ground floor of the new laboratory is laid out on the 'open-plan' system without dividing walls between the various sections. Ample window space and skylights provide excellent natural lighting. The upper floor consists entirely of office space, containing 15 offices in all. The ground floor houses the machinery and testing equipment of the pressure vessels, brittle fracture, and resistance welding sections, and the machine shop.

The pressure vessel section contains plant for testing vessels under pulsating and static pressure, and a high-pressure fatigue machine. The brittle-fracture section houses 700- and 3,000-ton test rigs for tensile tests, and a Charpy machine. A 4,000-ton test rig is being built which will be capable of producing fracture in mild steel plates 36 in. wide and 6 in. thick. The resistance-welding section contains medium and heavy machines for spot and projection welding and micro-welding devices for welding strain-gauge elements and electronic components.

Fracture in 6-in. mild steel plate

Construction has begun of a test rig with a capacity of 4,000 tons for tensile tests on welded joints in steel plate. Test rigs of 700 and 2,000 tons capacity have already been in use at BWRA for some time on plates up to 3 in. thick. The new machine is designed to produce fracture in alloy-steel plates up to 4 in. thick and mild steel plates up to 6 in. thick.

It will be constructed of weldable low-alloy steel of 25 ton/sq. in. yield and 6-in. plate thickness. Considerable use will be made of electroslag welding both in its construction and subsequently in the preparation of test specimens. As in the case of the 2,000-ton machine the well-established method of hydraulic capsule loading will be used, but with four large instead of 12 small capsules. Each of the four capsules will have longer working strokes and an operating pressure raised to 800 lb./sq. in. The apparatus has been designed to have a total weight of not more than 5 tons to suit existing lifting facilities. Test specimens will be welded to the load-bearing end beams of the machine and will be removed after testing by machine gas cutting.

The new test rig at BWRA is one of the most advanced of its kind and, as with the earlier machines, it was designed wholly by the Association staff. Plans have been sent on request to several overseas bodies, including the Paton Arc Welding Institute at Kiev. Tensile testing machines capable of fracturing very thick plate have particular application in nuclear energy engineering. Similar 4,000-ton test rigs are being built for the UKAEA and at Ghent University under the sponsorship of EURATOM.

Friction welding

Interest in friction welding was revived by the work carried out in the U.S.S.R. over the past two or three years. The process has been put to practical use in the Soviet Union and is being

applied in a number of production plants. The general belief is that the process is a Russian invention, but in fact British patents were taken out during 1941 and 1942, whereas the Russian patents were not registered until 1956.

The process is basically very simple. It consists of placing the ends of two bars in contact and rotating one of them, the frictional forces at the interface generating sufficient heat for welding to take place when the relative motion is stopped and sufficient force applied. Friction welding can be placed under the heading of pressure welding, the weld joint being made without the application of any external heat source. The process has some of the advantages of the more commonly known resistance-welding processes, but in addition requires much less power.

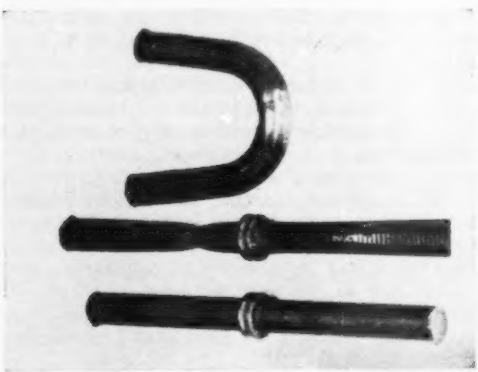
The machine recently demonstrated at BWRA was designed and manufactured by the central machine shop which forms the main service department for the establishment. Cost, manufacturing facilities and time available to some extent dictated the design. Again some aspects of the design have been influenced by the need for instrumentation with respect to existing recording equipment.

The machine (fig. 2) has an approximate maximum welding capacity of 1 sq. in. in mild steel and is powered by a standard 10 h.p. motor at 950 r.p.m. The axial force is provided by a 12-in. diameter double-acting air cylinder which with the 100 lb./sq. in. air line provides upset and friction forces to a maximum of 5 tons.

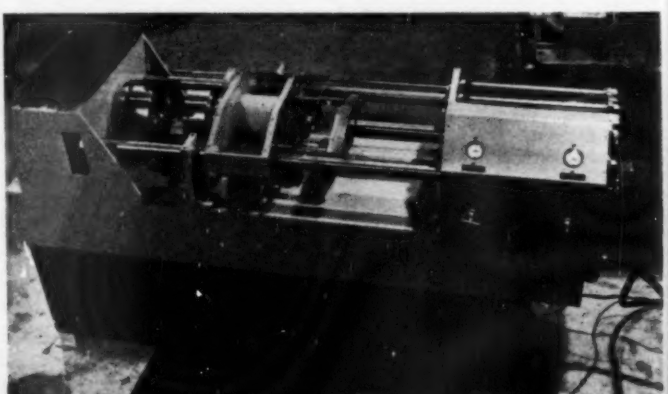
The drive is by power grip timing belt and stepped pulleys allowing speeds of 430, 640, 800 and 1,200 r.p.m. to be selected quickly and at a relatively low constructional cost.

The major mechanical problem involved is that of arresting the revolving head. The cost of doing this by a clutch, reversing the phases on the power supply to the motor or by capacitor injection are all costly and would have added several months to the construction time. The alternative solution was to allow the stationary shaft to accelerate up to the speed of the decelerating driven shaft which may also be further decelerated by a solenoid-operated brake (to be added later).

Release of the stationary spindle is accomplished by means of a simple torque arm on the non-driven shaft, the arm being held against rotation by a pivoted cantilever which will eventually carry strain gauges to enable torque to be recorded. The cantilever is held in the stop position by a solenoid-operated latch. The solenoid can be energized automatically at any pre-determined 'burn-off' distance. The solenoid control circuit also trips the motor and operates the air valves allowing the maximum force of the air cylinder to be applied to the work-piece. A simple beam is interposed between the cylinder push-rod and the moving



1 ABOVE *The efficiency of the friction-welding process is demonstrated in the photograph which shows welded ¾ in. dia. mild steel specimens after bending and tensile tests. An as-welded bar is shown at the foot of the illustration*



2 RIGHT *Prototype friction-welding machine constructed at BWRA laboratories. Friction welding is now an established process in the U.S.S.R.*

work head; this will eventually carry strain gauges so that friction and upset forces can be recorded.

Continuous current seam welding of stainless steels

Seam welding is used in many industries for producing continuous, pressure-tight welds. The basic principle of the process is that two sheets of metal are fed between a pair of copper electrode wheels while a high amperage low voltage current passes between them. The heating effect of the current is concentrated at the joint between the sheets and local fusion occurs.

A particular application is in fabricating parts of aircraft engines, and BWRA has been determining acceptable welding conditions for joining stainless steels for this purpose. The conventional way of seam welding is to switch the current on and off at very short, regular intervals to produce a series of overlapping welds. This produces satisfactory joints in most cases but 'interrupted' welds in some materials, especially highly alloyed stainless steels, are liable to crack. It has been found that the use of a continuous a.c. current for welding these materials eliminates the cracking, while it also appears that welding conditions are less critical than those for interrupted current. Strength tests and microscopic examination of the resulting welds are being carried out to confirm the suitability of the process (fig. 3).

3 *To determine the pressure tightness of continuous seam welds a 'pillow' test is used in which two plates, one equipped with an air nozzle, are welded together on all four sides and the resulting 'package' is inflated to failure. The specimen shown withstood a pressure in excess of* 1,400 *lb. sq. in.*

Measuring the temperature inside a spot weld

Although spot welding is used extensively for joining sheet metal components—it is the basis, for example, of practically all motor-car body constructions—little is generally understood of the way in which spot welds are formed. The process, while it can be accurately controlled and readily automated, remains more of a craft than a science.

When spot welding hardenable steels, the use of a simple weld pulse gives rise to joints which are brittle, and post-weld heat treatment is necessary to make the weld ductile. This is usually done in the machine but determination of the correct machine settings has been achieved, up to now, largely by trial and error. BWRA is therefore currently concerned with measuring the temperature cycle during spot welding by means of alumina thermocouples mounted either vertically inside the electrode or horizontally between the two sheets being welded. Very fine wire thermocouples are needed to follow the rapid changes in temperature during the welding cycle and the output is recorded on a high-speed ultra-violet galvanometer recorded or displayed on an oscilloscope.

To give an idea of the temperatures involved it has been found that alumina thermocouples can be melted inside the weld nugget, indicating a peak temperature of at least 2,000°C.

There are still some problems to be solved in using this technique, but once these are overcome the selection of spot welding and heat-treating conditions for specific materials should be greatly simplified.

Low-alloy steel pressure vessels

The continuing trend towards higher pressures and greater size of pressure vessels for modern plant lends importance to the need to investigate the use of stronger materials of construction. These would permit a reduction of plate thickness and weight over that required for vessels made in the traditional carbon-steel material, and thus allow further advances in operating conditions to be entertained.

Present-day design codes specify design stresses for low-alloy steels, which, for room-temperature service, are taken as a fraction of the tensile strength. However, these materials have a greater ratio of yield to ultimate stress than mild steel, and therefore greater exploitation of their advantages could be made if design were based on yield rather than ultimate strength.

The problem, which appears incapable of resolution on theoretical grounds, has been approached by BWRA through comparative tests on mild steel and low-alloy steel pressure vessels, under conditions simulating service. These tests should permit the selection of allowable stress on the basis

that vessels should have an equal margin of safety under their service conditions irrespective of material. The tests are being made on a series of identical spherical vessels, 30 in. diameter, $\frac{3}{8}$ in. thick, each vessel having a single 6-in. diameter opening. Resistance to overpressure, and failure following repeated applications of pressure, are being observed.

Electron microscope studies*

Properties and microstructure of mild-steel weld metal During the past year the electron microscope has been used extensively to study the relationship between the mechanical properties and the microstructure of mild-steel weld metal, both as-deposited and after reheating to different temperatures. It has been shown that, when as-deposited weld metal is normalized, the strength shows the same relationship with grain size and carbon content as in normalized wrought steels. However, as-deposited weld metal has a higher yield stress and yield-to-ultimate ratio than would be expected from the influence of grain size alone. This is particularly noticeable when the metal is deposited by processes resulting in high cooling rates. To avoid the ambiguity which must result from the study of a multi-run deposit, which contains in the same specimen weld metal in many different stages of heat treatment, recent work has been on single runs of weld metal. The results have shown that sub-critical annealing of up to 5 h. duration at 700°C. does not substantially affect the high yield strength which may even increase on reheating to temperatures between 600°C. and 700°C. Impact tests are also being carried out.

The microstructural investigation has had most interesting results. Metal arc weld deposits from electrodes of 6 S.W.G. or thinner have a structure of fine ferrite grains interspersed with thin layers of retained austenite. Fine non-metallic inclusions are visible but no carbides are seen until after reheating for a few hours at 250°C. Then very thin plates of carbide grow between the ferrite grains, and the retained austenite decomposes. Heating at higher sub-critical temperatures does not cause recrystallization of the ferrite even after 5 h. at 700°C., but the carbide structure changes to give spheroidal cementite particles at the grain boundaries. A fine etch structure of small surface bumps was a constant feature of the microstructure of both the as-deposited and reheated weld metal. No evidence of the growth of an intragranular precipitate was observed, although it had been suggested by other workers that this might be the explanation of the high strength.

The results tend to suggest that part of the weld metal strength is derived from a very stable substructure not revealed either by the light microscope or by the electron microscope using a replica technique. This substructure is only destroyed on heating into the austenite range. To investigate this, it is proposed to use the electron microscope to examine thin films of weld metal by direct transmission, which should reveal any substructure present. It is also proposed to investigate the effect on mechanical properties of heating weld metal to higher temperatures, testing at the elevated temperature and after cooling to room temperature.

Austenitic stainless steels The use of the electron microscope to determine the cause of heat-affected zone cracking during stress relieving of certain austenitic stainless steels has been very successful. It was found that cracking occurred when fine dispersions of a strain-induced precipitate formed within the grains of the heat-affected zone. The type of precipitate differed with different steels, the cracking severity being related to the density of the precipitate dispersion, which in turn depended on the amount of plastic strain taking place in the heat-affected zone and on the composition of the steel. In 18 Cr-13 Ni-1 Nb steels the precipitate was NbC, in titanium stabilized steel TiC, and in the unstabilized steel an $M_{23}C_6$ type carbide. No precipitate was observed within the grains of the 18 Cr-10 Ni-3 Mo type steel, and this was the only steel investigated which showed no tendency to heat-affected zone cracking. The observed effect of weld metal composition on cracking could be explained in terms of its hot strength. If this was high, the deformation in the heat-affected zone was correspondingly large and the result was a more highly embrittled structure, trying to accommodate a greater plastic strain.

The mechanism of failure is thought to be as follows. During stress relieving, relaxation of stress by creep is taking place in the heat-affected zone. In susceptible steels this is accompanied by precipitation of carbide on lattice defects introduced during deformation. This raises the strength of the grain interior and more of the overall deformation is taken by the boundaries sliding over each other. Stress concentrations are set up under these conditions, and they may become large enough to result in the nucleation of intergranular cracking.

Work is at present in hand on the mechanism of precipitation, studying thin films of metal in the electron microscope. It is hoped that the results will permit the fulfilment of the immediate objects of the investigation:

(*a*) To define the conditions under which existing steels can be used.

(*b*) To understand the mechanism of embrittlement and thus predict chemical compositions which will combine a good creep strength with acceptable weldability.

*Report of the Council, 1960-1961, BWRA.

Ultra-high-vacuum techniques

OVER THE PAST TWO YEARS there has been a growing demand for high-vacuum equipment capable of producing pressures below 1×10^{-9} Torr. These pressures have been found necessary in order to produce high-purity thin films of various materials by evaporation techniques. The main application of these films, at present, is in the field of solid-state physics and semi-conductors where studies of the magnetic and other properties of such films are carried out.

Previous apparatus designed for this low-pressure work has involved the development of special vacuum seals in order to achieve a negligible leak rate to enable the vacuum pumps and liquid nitrogen traps to evacuate the work chamber to the required ultimate pressure. The development of such seals, involving careful leak testing, becomes a somewhat uneconomic process.

One method is to use all glass bakeable systems, but these are essentially laboratory tools and can only deal with very small quantities of materials to be evaporated. In addition, the available ultimate pumping speed of such a system is very low, which means that any outgassing process of the vapour source can take four or five days or longer.

Single chamber systems, with metal gasket seals of various types, bakeable up to approximately 400°C. have been built, but these systems generally require upwards of 12 h. to produce the ultra-high-vacuum conditions necessary to evaporate materials.

Design principles

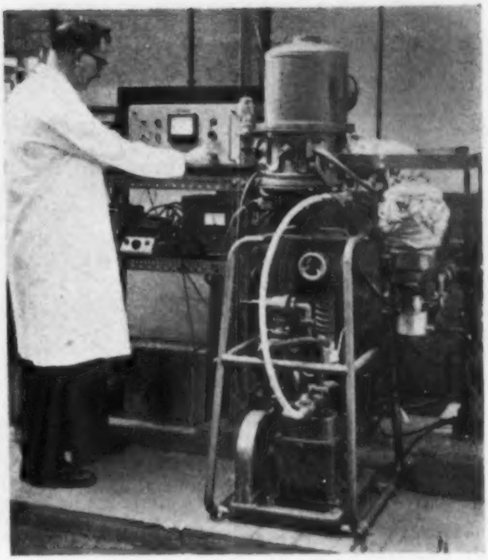
An ultra-high-vacuum coating plant must satisfy the following conditions:

(1) Short pump down times and process cycles.
(2) Easy loading and unloading of the UHV work chamber with guaranteed attainment of the low ultimate pressure after reloading with fresh material.
(3) Use of standard interchangeable components throughout the system.

These conditions have been satisfied in the design of equipment demonstrated recently at the Boreham Wood factory of Leybold-Elliott Ltd., a member of the Elliott-Automation Group.

The laboratory prototype (fig. 1) was developed by Dr. Ehlers and Dr. Moll in Cologne to test the design principle of using a double chamber system.

Since leak rate is a function of pressure difference across any seal, if the outside pressure is reduced to a suitable value, the potential leak rate is reduced to a negligible quantity. Thus in this plant the interspace enclosed by the outer chamber is evacuated by a perfectly normal type of pump set, achieving an ultimate pressure of the order of 10^{-5} Torr.



1 *The laboratory research ultra-high-vacuum plant at the Boreham Wood factory of Leybold-Elliott Ltd.*

The mild-steel outer chamber has electrode entries sealed by conventional synthetic rubber O rings or gaskets, the chamber flange bearing against a trapezoidal-type washer.

The inner UHV chamber, of stainless steel, has its seals of copper, with the flange seal merely clamped. Lead-throughs are provided, extending from the power supplies through the outer chamber entries to the copper gasket sealed entries to the UHV chamber. Windows are provided in both chambers to enable operators to observe the process or experiment being carried out. Similarly, rotary transmissions are provided to move masks or shutters over the vapour source during outgassing processes.

In order to produce clean, low-pressure conditions the inner chamber is outgassed by resistance heating to a temperature of 400°C. or 450°C. A liquid air trap of the Meisener type is fitted inside the UHV enclosure, which is pumped by a fractional oil diffusion pump of speed 120 l./sec. This is backed by the other diffusion pump and a double-stage gas ballast rotary vane pump of displacement 5 m.^{3}/h.

continued on page 338

General view of the laboratory

Heat-treatment research at A.E.I.-Birlec

THE BIRLEC Research and Development Department came into being nearly 25 years ago, soon after the formation of the company. For the greater part of its existence it has been located at the company's main premises in Tyburn Road, Birmingham. In recognition of the growing importance and scope of its work, the department was transferred to the present larger premises at Wood Lane in 1960.

The new laboratory is housed in a building with a total floor area of 20,000 sq. ft. It is believed that this is the largest and best equipped research and development laboratory in the furnace industry. The department employs 30 people, about 50% of these having technical qualifications in physics, metallurgy, chemical engineering, mechanical engineering or electrical engineering. Four major sections deal respectively with metallurgical processes, furnace design, gas plants and dryers, and the testing of prototype equipment. A certain amount of work is also undertaken for an associate company, Birlec-Efco (Melting) Ltd., and the staff is responsible for operating a pilot smelting plant for this company near Aldridge.

Facilities

Electric power up to a rating of 750 kVA. is distributed throughout the department, as well as town's gas up to 9,000 cu. ft./h., mains water up to 4,000 g./h., recirculated cooling water up to 2,000 g./h., and compressed air up to 18,000 cu. ft./h. An overhead crane of 10 tons capacity with a 24-ft. lift serves the entire area.

Rooms are provided for the preparation of metallurgical specimens, microscope and photographic work. Instruments available include infrared gas analyser, Hersch oxygen analyser, Foxboro dewpoint measuring and control units, and a 'moisture monitor' dewpoint instrument. A Vickers projection microscope and a wide range of temperature measurement and control instruments are also available.

Investigations in progress

The scope of the department's programme is the investigation and development of all types of equipment manufactured by the company. Work includes examination of basic design principles, the testing of specific components, the performance

evaluation of prototype plant, and the testing and final adjustment of specialized production equipment to provide data for future design and for the instruction of field engineers. The following account gives brief details of some specific research projects related to the heat-treatment field.

Bright annealing stainless steel strip The bright-annealing process is carried out in an atmosphere of cracked ammonia, and conventional annealing equipment incorporates a heat-resisting nickel-chrome alloy muffle to exclude contamination of the gas. The muffle is expensive and for many applications needs to be replaced from time to time. In addition, the heating elements are necessarily fitted outside the muffle and are thus subject to heavy duty.

In an experimental furnace the metal muffle has been eliminated by lining the furnace chamber with a refractory which has been selected for its ability to remain in equilibrium with reducing atmospheres of low dewpoint at high temperatures. By this means, the disadvantages characteristic of the conventional equipment are overcome. The furnace incorporates strip-handling gear and experiments can be undertaken with strip up to 6 in. wide. Work has been carried out with austenitic, ferritic and martensitic stainless steels with completely satisfactory results in all cases. Extension of the new principle to other bright-annealing applications is receiving consideration.

Vertical sealed-quench furnace Installed in the laboratory is a full-scale sealed-quench carburizing unit which includes the usual quenching facilities but incorporates a number of features to simplify the heating and quenching system and associated handling of work through the furnace. In addition it has been utilized as a means of subjecting unprotected heating elements of various designs to a process of life-testing under conditions closely approximating those obtaining in gas-carburizing practice.

Prototype shaker-hearth furnace A full-scale furnace is operating in the laboratory. This is a heavy-duty unit which is intended to extend the scope of continuous-hardening equipment up to the range of the larger cast-link conveyor type of furnace. The principal feature is a very much modified hearth design intended to reduce hearth distortion.

Experimental vacuum furnace This is a small high-vacuum high-temperature unit, used to provide both process and design data. Applications include annealing, stress relieving, brazing and sintering.

Radiant tubes

Test rig for oil-fired radiant tubes A 15-ft. oil-fired radiant tube for furnace heating is operating in a specially constructed test rig. Objects of this work include the investigation of factors affecting temperature distribution along the length of the tube, the identification of operating conditions likely to lead to soot deposition in the tube, the design of a burner suited to radiant tube operation and the collection of operating data.

Single-ended gas-fired radiant tube A radiant tube in which a proportion of the products of combustion is recirculated within the tube is undergoing development. The objects of this work are reduction of temperature gradients along the length of the tube and the development of a gas-fired unit which is interchangeable with the standard Birlec electric radiant tube-heating element.

Electric radiant tubes Two experimental electric radiant tubes are operating in the laboratory, each incorporating novel features intended to reduce first cost, improve operating characteristics and increase working life.

Gas plant

Prototype exothermic gas plant with flame failure protection This development represents an attempt to reduce both cost and size of exothermic gas generators and to improve operation. A gas/air mixture is combusted in a tunnel-type burner and the product gas is cooled by being passed through

Development of a vertical pusher annealing furnace

a curtain of water. A plant capable of delivering 5,000 cu. ft./h. of exothermic gas is at present working in the laboratory.

The objects of the investigation in progress include the collection of data regarding the design and operation of the burner, demonstration of the efficiency of the direct water-cooling unit, and assessment of the quality of the product gas with regard to metallurgical applications.

A number of commercially available flame failure protection devices has been tested on the laboratory plant, with a view to selecting the unit best suited for incorporation in the production generator.

Removal of carbon dioxide from the products of combustion of a gas/air mixture The company already markets a standard range of stripping units of the type in which carbon dioxide is absorbed in an aqueous solution of mono-ethanolamine, but work is in progress on the development of two possible alternative processes.

The first of these involves the use of a suitable adsorbent material and is similar in operating principle to the conventional gas-drying equipment containing activated alumina. This represents a considerable simplification in stripper design and operation when compared with the more usual mono-ethanolamine type unit. The equipment consists of a large rig in which the capacity of the adsorbent has been determined in relation to conditions of operation and reactivation.

The second process under investigation is that of the absorption of carbon dioxide in water at high pressure. An experimental unit has been erected and tests should commence in the near future.

Improved catalysts for endothermic-type gas generators Work is in progress on the development of improved catalysts for use in this type of plant. Experiments are carried out in a full-scale Birlec endothermic-type gas generator of conventional design.

Adsorbent-type dryers for compressed air and gases A new series of dryers in this class has recently been marketed. The essential design data and final plant development and type testing were carried out in the laboratory. Examples of the smaller sizes of dryers in this range have been retained in the laboratory for experimental purposes.

Air-conditioning units Work is in progress on the further development of the dehumidifying unit. Equipment has been constructed which simulates a section of standard dryer and experiments are planned in which the behaviour of different desiccants in this type of dryer will be studied.

Other investigations in progress

The rate at which oxygen will diffuse into a controlled atmosphere furnace against an outward flow of gas has recently been studied. Experiments demonstrating the validity of the theoretical predictions are now proceeding.

A bench-scale investigation is being initiated in which the high-temperature carburizing characteristics of various steels will be studied. This work is also intended to yield process data relating to high-temperature carburizing.

Experiments are being made to develop seals suitable for use on the ends of a high-vacuum continuous-strip annealing furnace.

The assessment of the properties of desiccants on a comparative basis is a subject of considerable interest to the company. An appropriate technique and apparatus have been developed and are now in routine use in the laboratory.

Ultra-high-vacuum techniques

concluded from page 335

These diffusion pumps are standard models with water-cooled baffles fitted to the intake ports. The UHV system has a bakeable manifold with a liquid nitrogen trap fitted above the water-cooled baffle.

This system is capable of achieving a pressure of 10^{-9} Torr in 4 to 5 h. with ample residual capacity to maintain a low pressure during evaporation and outgassing of the vapour source.

Sequence of operations

(1) Rough evacuation of the chambers for 10 min.

(2) Fore diffusion pump started. This takes about 20 min. to reach a pressure of 10^{-4} Torr.

(3) Inner chamber and manifold outgassed for $1\frac{1}{2}$ h. (450°C. for initial operations).

(4) Second diffusion pump started during outgassing.

(5) Inner chamber cooled by air for 10 to 15 min.

(6) Liquid nitrogen injected into the traps. This takes about 15 min.

The inner and outer chamber pressures are measured by Alpert-type thermionic ionization gauges. The control unit for the UHV gauge is connected to a high gain amplifier in order to accurately measure pressures below 10^{-8} Torr.

Thus with an outgassed plant a pressure of 10^{-8} Torr can be reached within 3 h.

The unit exhibited and shown in fig. 1 is the laboratory research plant. Models based on this principle, the UPO5 and UPO6, are available as standard plant and incorporate a larger number of facilities for coating work.

NEWS

United Steel purchase Barrow Steel Works

THE United Steel Companies Ltd. have purchased Barrow Steel Works Ltd. from the Iron and Steel Holding and Realization Agency for the sum of £2,200,000.

United Steel has managed this works since 1943 for a token remuneration. As a conventional steel-producing unit, the works has not been economic throughout this period. The re-rolling mills are not in themselves uneconomic, but their steel requirements are insufficient to support their own steelmaking furnaces, while the location of Barrow prevents the regular acquisition of billets from other sources of supply.

Continuous casting was considered to be the most promising form of small-scale billet production, based on local scrap supplies, and in the hope of converting the works into a profitable undertaking a pilot continuous casting unit was installed in 1952. Since 1954, United Steel has borne the operating cost of the subsequent experimental work in view of its own interest in continuous casting developments.

The process is now working successfully, and two full-scale twin-strand continuous casting machines, together with a 20-ton electric arc furnace, are being installed. This plant is expected to come into production in the near future, and should provide billets at a cost which will enable the Barrow re-rolling mills to operate at a profit and so justify their continued existence.

New cold-forging factory in Germany

The Camp Bird Group of companies announces the opening and start of production of a new £500,000 factory in West Germany, for cold-forging plant. It is a member of the International Cold Forging Group whose factories are situated at Ravensburg (Germany), Monaco, and at Sunbury-on-Thames in the U.K. The new factory is located at Lockweiler between Saarbrücken and Trier, and will be known as Saarländische Werkzeug und Maschinenfabrik, Walther Nothelfer G.m.b.H.

Apart from the manufacture of cold-forging equipment for direct sale, it is also planned to maintain a complete cold-forging installation in production as this will enable customers' plant to be proved under actual operating conditions.

The works comprise two ground-level buildings of 40 x 180 ft. and 177 x 184 ft. respectively. A third single-storey building is nearing completion. Considerable ground around the factory has been acquired for the expected future expansion.

International research corporation starts operations in U.K.

A new corporation to provide research, technical information, investment advisory, and management services in the United States and Europe has been organized by Dr. Clyde Williams, Columbus, Ohio, and his American and European business associates.

Dr. Williams has been a frequent visitor to London between 1951 and 1956 when, as president of Battelle Memorial Institute, he established Battelle's European operations which consisted of research institutes in Frankfurt and Geneva, and technical offices in London (at John Adam House, John Adam Street, London, W.C.2), Paris, Milan and Madrid. In 1956, Dr. Williams discussed plans for building a laboratory in Britain with Mr. Harold Macmillan (then Chancellor of the Exchequer), Mr. Peter Thorneycroft (President of the Board of Trade), and Sir David Eccles (Minister of Education). Incorporated in Ohio, the Clyde Williams Corporation has its main offices at 50 West Gay Street, Columbus; branch offices will be maintained in London and Paris. Dr. Clyde Williams, Dr. F. R. Hensel, Dr. L. Kermit Herndon, and Mr. H. H. Jackson have been appointed to the board of directors. Officers of the Corporation will be: Dr. Williams, president and chairman; Mr. Jackson, executive vice-president; Dr. H. E. Zentler Gordon, vice-president; Mr. Clyde Williams, Jnr., secretary and treasurer, and others to be named.

'A major purpose of the company is to provide liaison between European and American technology,' Dr. Williams stated in announcing the launching of the new

GKN House—the new London office of the Guest Keen and Nettlefolds Group of companies—stands on the site of the old Stoll Theatre at 22 Kingsway. It is a seven-storey building with 92,000 sq. ft. of office accommodation furnished and equipped in modern style. The building has two basements for car parking with space for 46 cars. On the ground floor are showrooms, where a selection of products manufactured by the Group is displayed. Adjacent to the showrooms is a cinema and lecture theatre accommodating 50 people. Furniture throughout the building is supplied by Sankey-Sheldon Ltd., different styles being adopted for different types of office

enterprise. ' What we are doing is to hasten the process of across-the-Atlantic information dissemination as it relates to technology, to markets for products, and to finance and investments.'

Dr. Williams, who introduced the concept of contract research in Europe with his establishment of the laboratories of Battelle Institute in Frankfurt, Germany and Geneva, Switzerland, approximately ten years ago, said that among other things the new firm is advising American companies where to place research projects. For this purpose, European nationals highly trained in research management will be attached to the London and Paris offices.

Official opening of Solartron headquarters

Solartron instruments, specialized equipment and various machines and systems were demonstrated and exhibited at the opening of the headquarters of the Solartron Electronic Group Ltd., Farnborough, Hampshire, by the Rt. Hon. Reginald Maudling, M.P., President of the Board of Trade, last July.

With the completion of the second phase of the five-year building programme of the Solartron Electronic Group Ltd., the offices, research, production and ancillary buildings, such as the canteens, occupy 175,000 sq. ft. of working floor area. The first phase consisted of 50,000 sq. ft. of workshop area and a one-storey administrative section attached. The second phase consists of a four-storey block of 35,000 sq. ft. and 70,000 sq. ft. of production floorspace.

COMPANY NEWS

National-Kayser

The National Machinery Co. announces the amicable termination of its agency agreement with Buck & Hickman, effective July 1. Henceforth the National Machinery Co. will act directly through an office already established in the U.K. The agency which its affiliate, J. G. Kayser, has had with the Ed Brand Co. of London has also terminated.

The reasons for the above reflect the changes taking place in present-day business. The new factory in Nurnberg, Germany, means that machine demonstrations, technical conferences and assistance are only a few hours away. In addition, National-Kayser services in the U.K. have grown to the point where continual technical contract with clients is the rule.

Full use of the facilities in Sutton Coldfield, in Tiffin, Ohio, U.S.A., and at the Nurnberg works is welcomed. The address of National-Kayser is 429 Birmingham Road, Sutton Coldfield, Warwickshire (ERDington 4192, Birmingham).

Vitreous Enamellers join Promotional Council

The Vitreous Enamellers' Association, for 25 years the industry's trade association, has joined forces with the Vitreous Enamel Development Council, the industry's promotional body. Its members, in the association's new role, will form a Jobbing, Signs and General Division of the Council.

Microwave Instruments Ltd.

The directors of Hilger & Watts Ltd. announce that their company has acquired the whole of the issued share capital of Microwave Instruments Ltd.

Mr. J. Bilbrough, A.M.BRIT.I.R.E., will continue to be managing director of Microwave Instruments Ltd., whose factory is at North Shields in Northumberland. They are precision and electronic engineers, well known for their wave-guide components and microwave test equipment, which are complementary to those for millimetre wavelengths manufactured by Hilger & Watts. The two companies have been collaborating for over a year in the development of measuring apparatus for electron spin resonance.

Wray (Optical Works) Ltd.

Hilger & Watts Ltd. announces that Wray (Optical Works) Ltd. has come into association with the company, and feels that thereby the optical instrument industry of this country will be strengthened. Mr. A. W. Smith (a former president of S.I.M.A.) is continuing as managing director of Wray (Optical Works) Ltd.

Welding Equipment Division formed by English Electric

The welding business of English Electric is being co-ordinated in a newly-formed Welding Equipment Division which is being set up at the company's Accrington works.

The manager of the division will be Mr. R. H. Boughton; the chief engineer, Mr. F. Mullery; the sales manager, Mr. E. H. Ayres.

Edward G. Herbert Ltd.

At the last a.g.m. of Edward G. Herbert Ltd. Mr. Alan Kiernan, M.I.MECH.E., and Mr. Sam Smiley, F.C.C.S., A.A.C.C.A., retired from the service of the company.

Mr. Chas. E. Rogerson, O.B.E., F.C.A., and Mr. V. M. Marshall, M.I.LOCO.E., were appointed directors and the board later elected Mr. Chas. E. Rogerson vice-chairman.

LECTURES AND COURSES

Dr. D. G. Christopherson, O.B.E., F.R.S., Pro-Vice-Chancellor of the University of Durham and Warden of the Durham Colleges, has accepted an invitation to give the first annual lecture of the British Conference on Automation and Computation (BCAC).

He will speak on the subject, ' Mathematics—Friend or Foe?,' at the lecture theatre of the Institution of Electrical Engineers, Savoy Place, London, W.C.2, on Wednesday, September 27, at 5.30 p.m. The lecture is open to all members of the 31 member societies of BCAC; others should apply for tickets (free) from the honorary secretary, BCAC, c/o The Institution of Electrical Engineers, Savoy Place, London, W.C.2. Tea will be served at 5 p.m.

' Hydrogen in steel '

BISRA, who are organizing a conference at Harrogate from October 11 to 13, announces that **Professor A. R. Troiano,** of the Case Institute of Technology, Cleveland, U.S.A., will give the opening lecture. He will deal with the role of hydrogen in the mechanical behaviour of metals.

Full details of the arrangements, together with application forms, can be obtained from the Technical Secretary, Metallurgy Division, BISRA, 11 Park Lane, London, W.1.

The Institution of Production Engineers are holding a summer school at the College of Aeronautics, Cranfield, August 29 to September 1.

The subject is ' The interrelation of work study, ergonomics, operational research and cybernetics and their application to production engineering.' Early application is advisable as the number of places is limited to 120. The inclusive fee for the school, which will also cover accommodation, meals and gratuities, is 10 guineas.

Change of address

The new address of John Miles & Partners (London) Ltd., consulting engineers, is Moor House, London Wall, London, E.C.2 (METropolitan 0471).

PEOPLE

THE ADMINISTRATORS of the Sir George Beilby Memorial Fund, representing the Royal Institute of Chemistry, the Society of Chemical Industry and the Institute of Metals, have decided to make awards from the Fund in 1961—each consisting of the newly instituted gold medal with a prize of 100 guineas—to the following:

Dr. Constantin Edeleanu, M.A., PH.D.—in recognition of his work on the corrosion of metals and alloys, with special reference to the development of the potentiostat technique and its applications to the study of practical problems, and on the characteristics of corrosion reactions in fused salts.

Prof. Jack Nutting, M.A., B.SC., PH.D., F.I.M.—in recognition of his work in physical metallurgy, especially in the application of the electron microscope to the study of the relationship between microstructure and mechanical properties of metals and alloys and to the investigation of phase changes and dislocation interactions.

Dr. Constantin Edeleanu was born in Rumania, and came to Britain in 1937. He was a pupil at the Morrison Academy, Crieff, and then went up to Cambridge in 1941. After a brief period in the Army he returned to Cambridge in 1945. In 1946 he joined Dr. U. R. Evans, working on the stress corrosion of light alloys. In the course of that work he became interested in pitting reactions and developed the autocatalytic mechanism for pitting first proposed by Dr. Hoar. A somewhat surprising conclusion of the work was that the true rate of corrosion during the pitting of aluminium in a given solution was a constant. Direct experimental proof for this was, however, only obtained some ten years later.

In 1950 he joined the Brown Firth Research Laboratories, where he was in charge of a section dealing with long-term alloy development. Stress corrosion, especially in high-pressure water systems and nuclear-reactor cooling systems, was of particular interest, and the work demonstrated that failures in such systems were, as a rule, associated with concentration effects arising either during heat transfer or in heating and cooling cycles.

His contribution to the development of corrosion-resistant steels was twofold. Firstly, it was shown that a detailed knowledge of the structure of alloys was necessary in order to make progress, and secondly, the work helped to establish the potentiostat techniques for corrosion studies. This work also showed that the potentiostat was a powerful metallographic tool but, perhaps more important, it showed that it was possible to stabilize passivity to a remarkable extent by 'anodic protection' and that this could form a basis for a practical method of protection in chemical plant. The process is now being used on a fairly large scale in the United States.

In 1956 he joined the Tube Investments Research Laboratory at Hinxton, to form a corrosion group. At first the work consisted mainly of studying the factors that control corrosion in some fused-salt systems of interest to the nuclear engineering industry. One of the main results was to show that such systems could be treated in a somewhat analogous way to aqueous systems but that there were important differences. For instance, it was not possible to neglect solid-state diffusion and also the vapour pressures of some of the compounds had to be considered. Graphical methods of representing fused electrolyte-metal systems were developed, and this approach is now being used to see whether it might also elucidate certain problems in extraction metallurgy.

More recently Dr. Edeleanu has returned to stress corrosion and, in conjunction with Dr. A. J. Forty, has attempted to interpret this type of failure in a more physical manner. This led naturally to the question of the part played by dislocations and other such structural features in corrosion. With G. A. Bassett a start was made in this field, techniques such as transmission electron microscopy having played a useful part.

Recent electrochemical studies have included work on passivity, with particular reference to anodic protection.

Prof. Jack Nutting was educated at Mirfield Grammar School, Yorkshire, and the University of Leeds where he graduated with first-class honours in metallurgy in 1945. He was then appointed research assistant to the late Professor Preece at Leeds and was awarded the degree of Ph.D. in 1948 for a dissertation on the overheating and burning of steel. Subsequently he was employed by the British Iron & Steel Research Association and was seconded to the Cavendish Laboratory, Cambridge, where he worked with Dr. V. E. Cosslett and Professor E. Orowan, developing techniques for the examination of metals with the electron microscope. In 1949 he was appointed demonstrator and in 1954 lecturer in the Department of Metallurgy, the University of Cambridge—an appointment he held until 1960 when he returned to the University of Leeds as professor of metallurgy in the Houldsworth School of Applied Science.

During his work at the Cavendish Laboratory he became convinced that the electron microscope could be more widely applied to problems in metallography. On transferring to the Metallurgy Department in Cambridge he had the opportunity of testing his convictions, and together with a group of research students a wide variety of metallographic problems were investigated. The first topic to be studied was the microstructural changes occurring during the tempering of plain carbon steels. With the development of improved replica techniques, the work on tempering was extended to alloy steels; in particular the phenomenon of secondary hardening in ferritic steels was investigated, whilst attempts were made to correlate the microstructure of these steels with their creep behaviour. The techniques developed for ferritic steels have been modified subsequently for application to austenitic steels, and again the results obtained have been correlated with creep behaviour.

The carbon-extraction replica methods, which had proved successful for studying steels, were completely unsatisfactory when applied to aluminium alloys. In order to examine these materials entirely new specimen preparation methods, thin-foil techniques, were developed so that a metal could be examined directly by transmission in the electron microscope. With these techniques it has been possible to detect Preston-Guinier zones in a variety of aluminium alloys and also to detect coherency strain fields around precipitated phases. Further modifications of the thin foil technique allowed specimens to be prepared from bulk samples, and it became possible to study phase transformations involving shear. The martensite reaction in plain carbon and alloy steels was examined, and the plate-like form of martensite, previously thought to be homogeneous, was found to be composed of a stack of fine twins.

Investigations on the way two-phase alloys deformed plastically were begun with the aid of replica techniques, but when other workers showed that dislocations and dislocation interactions could be studied by examining thin foils in the electron microscope it became possible to determine directly the interactions between dislocations and precipitates. This type of investigation is now being extended to the study of precipitate dislocation interactions during creep.

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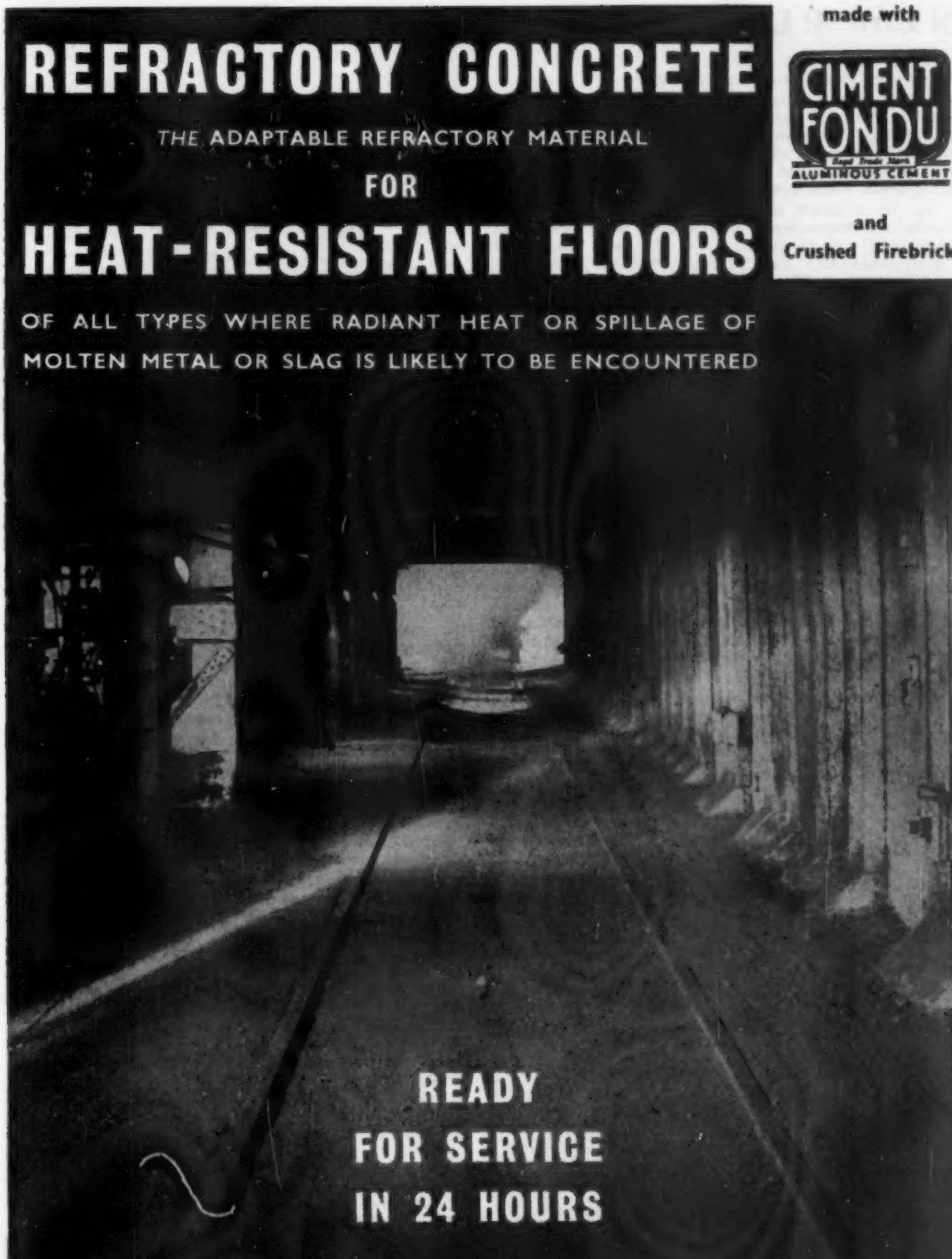

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NEW PLANT

High-precision laboratory furnaces

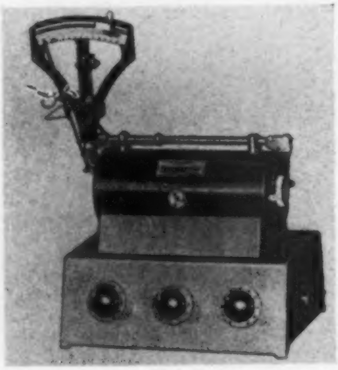
A RANGE of precision laboratory furnaces with exceptionally high stability of temperature (±0·25°C.), an interchangeable drum control for thermal cycling, and facilities for heat-loss compensation at the ends of the tube, have just been introduced by Shandon Scientific Co. Ltd., 6 Cromwell Place, London, S.W.7.

In the standard range there are three models for maximum temperatures of 1,050°C. with tube sizes respectively of 35 mm. i.d. × 290 mm. length (for operation on 115 V. a.c.), 60 mm. × 450 mm. (220 V. a.c.) and 75 mm. × 700 mm. (220 V. a.c.); and two models for maximum temperatures of 1,250°C., with tube sizes of 35 mm. × 290 mm. (115 V. a.c.) and 60 mm. × 450 mm. (220 V. a.c.). Two special models for maximum temperatures of 1,500°C. are also available, with tube sizes of 16 mm. × 250 mm. (115 V. a.c.) and 35 mm. × 400 mm. (220 V. a.c.).

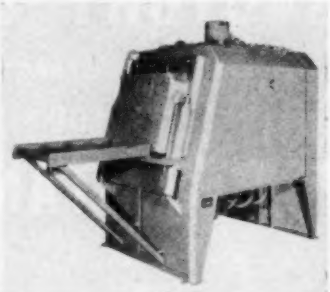
The furnaces are of versatile design. The standard models may be controlled either by a thermostatic regulator or a chronograph drum control for thermal cycling, these temperature-control arrangements being easily interchangeable by the user.

The thermostatic regulator is a reliable mechanism enabling the furnace temperature to be controlled within ±0·25°C. at any desired level below the maximum. The device consists of a graduated scale and a pointer operated by a dilatable wire running through the furnace body. The scale is fitted with a stop which may be set to any desired temperature. When the pointer reaches the stop, the current is automatically cut off from the windings via the 8-V. relay. The temperature is set initially with the aid of a thermocouple.

The thermostatic regulator is interchangeable on the standard models with a chronograph drum control for thermal cycling. The drum is clockwork-operated as standard, but an electrically driven version is available as an alternative. The drum carries a temperature chart on which the desired thermal cycling programme is marked out by a strip of thick cardboard. In operation, the drum revolves at a set speed (the gearing arrange ments provide two speeds) and, as the furnace heats up, a specially designed contact on an arm connected to the dilatable wire moves across the drum. When it meets the cardboard, the relay operates and the current is cut off. In this way the contact follows closely the edge of the cardboard strip, alternately making and breaking the circuit to the windings, and thus holding the temperature to the planned programme.



1 *Precision laboratory furnace*



2 *Ipsen controlled atmosphere furnace*

The three separate windings are mounted side by side along the tube and may be controlled individually by shunt rheostats, thus enabling the temperature to be varied along the tube for special work or to compensate for the extra heat loss at the tube ends inherent in this type of furnace.

All models may be mounted vertically or horizontally and may be fitted with a water jacket, *e.g.* for use in a glove box.

The two special 1,500°C. models are developed from the standard types. They are available only with a chronograph drum control, which is not interchangeable with a thermostatic regulator, and incorporate special windings of platinum-rhodium wire. Like the standard models they may be mounted either vertically or horizontally on any of the three available bases.

Controlled atmosphere box furnaces

The manufacture of a complete line of controlled atmosphere box furnaces is announced by Ipsen Industries Inc., Rockford, Illinois. Operating temperatures of 1,315°C. can be maintained in the smaller units, and larger units will be capable of temperatures to 1,230°C.

The furnaces are available as gas-fired, oil-fired or electrically heated units. The units maintain excellent temperature uniformity, since the charge is heated by 100% forced convection. The charging door and door plate of the tightly sealed units are of one-piece normalized cast meehanite. These castings are rigid and resist warping. Sliding surfaces are machined flat and the door is raised by dual airdraulic cylinders.

Higher temperatures can be obtained because there is no alloy inside the furnace. The protected atmosphere is circulated by a ceramic fan capable of driving the atmosphere at a speed of 2,500 ft./min. Heating is accomplished by means of the Ipsen super-alloyed ceramic heating tubes that are impervious to high-carbon and high-hydrogen atmospheres as well as resistant to the extreme changes of either heating or cooling. An exclusive 100% premix ceramic burner fires upward to assure complete combustion within the heating portion of the tube. The ceramic burner has extremely long life, since it is not affected by high-temperature operation or exposure to contaminating-type atmospheres. Coupled with the use of Ipsen super-alloyed ceramic flame busters, this feature increases heating efficiency, resulting in more

heat transferred to the work from a given amount of fuel.

Ipsen box furnaces can be manually charged, or they can be used with mechanized loaders and unloaders for production-line heat-treating operations. Development of box-type furnaces is expected to widen the complete line of heat-treating equipment offered by Ipsen Industries Inc. These units are expected to find uses in tool room applications, and for other applications where controlled atmosphere heating is desired, but where it is unnecessary to hold work under conrolled atmospheres following the heating operation

Coatings for shot-blasted steel

Camrex Paints Ltd. have recently introduced on to the market two new coatings—Camrex Shot Prime and Camrex Shot Kote. Both of these preparations are for application over newly shot-blasted iron and steel surfaces.

Both preparations have been developed for spray application. It is, therefore, possible to shot-blast very large areas without interruption, dust down and apply a spray coat. These preparations dry within three minutes and present a surface which will not corrode for a considerable period. The surface so prepared is resistant to rust for approximately 12 months and is capable of receiving any type of paint. Adhesion to this surface is excellent. The covering capacity is approximately 60 sq. yd./gal. with a film thickness of 1–1½ thou. The finish is aluminium and the material itself is non-toxic. Plates treated with these materials may be welded without any trouble and, in fact, both materials have been tested by various shipyards for this characteristic. In addition, welding on one side of the plate does not affect the protective quality of the Shot Prime on the other side, if both sides have been shot-blasted and treated. No toxic fumes are evolved on welding.

Classified Advertisements

FIFTEEN WORDS 7*s*. 6*d*. (minimum charge) and 4*d*. per word thereafter. Box number 2*s*. 6*d*. including postage of replies. Situations Wanted 2*d*. per word.

Replies addressed to Box Numbers are to be sent, clearly marked, to Metal Treatment and Drop Forging, John Adam House, John Adam Street, London, W.C.2.

MACHINERY WANTED

CLIPPING PRESSES—one 350/400 ton, one 600 ton. To be of recent manufacture and in good condition. Clydesdale Stamping Co. Ltd., Netherton, Dudley, Worcs.

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COVENTRY CLIMAX AND CONVEYANCER FORKLIFT TRUCKS!!! A choice selection of electric, petrol and diesel models. From 1,500 lb. to 5,000 lb. capacity; 9 ft. to 12 ft. lifts Full details, photograph and prices from Speed Electrics, Dept. MTDF, Church Street, Basford, Nottingham. Tel. 75716.

TWO—HEENAN & FROUDE TYPE P.66 OIL-COOLER UNITS, complete in all details; condition unused. Two—Vickers-Detroit Hydraulic Pumps, Type V.105.C.—Whitefield Machinery & Plant Ltd., 48 Chatham Street, Edgeley, Stockport.

MACHINERY FOR SALE—continued

'ALLDAYS & ONIONS' 2-CWT. PNEUMATIC POWER HAMMER for sale. Overhanging type without slides. Length of stroke 14". Size of bar worked 4" diameter. Diameter of ram 9¼". Arranged motor drive 440/3/50. Complete with anvil. Weight about 50 cwt. F. J. Edwards Limited, 359 Euston Road, London, N.W.1 or 41 Water Street, Birmingham 3.

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500	B.E.T.	11 kV.–400/230
500	Parsons	6·6 kV.–3·3 kV.
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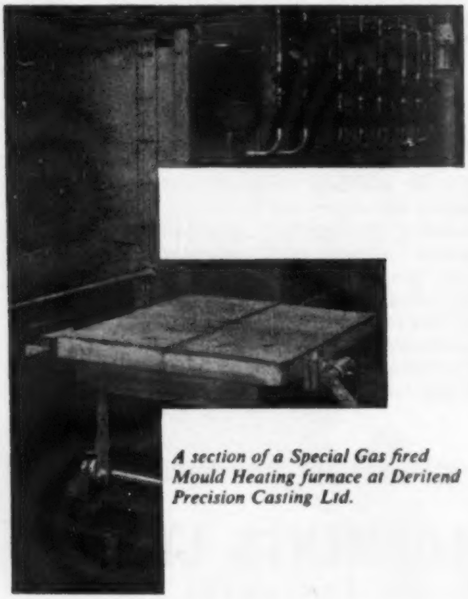

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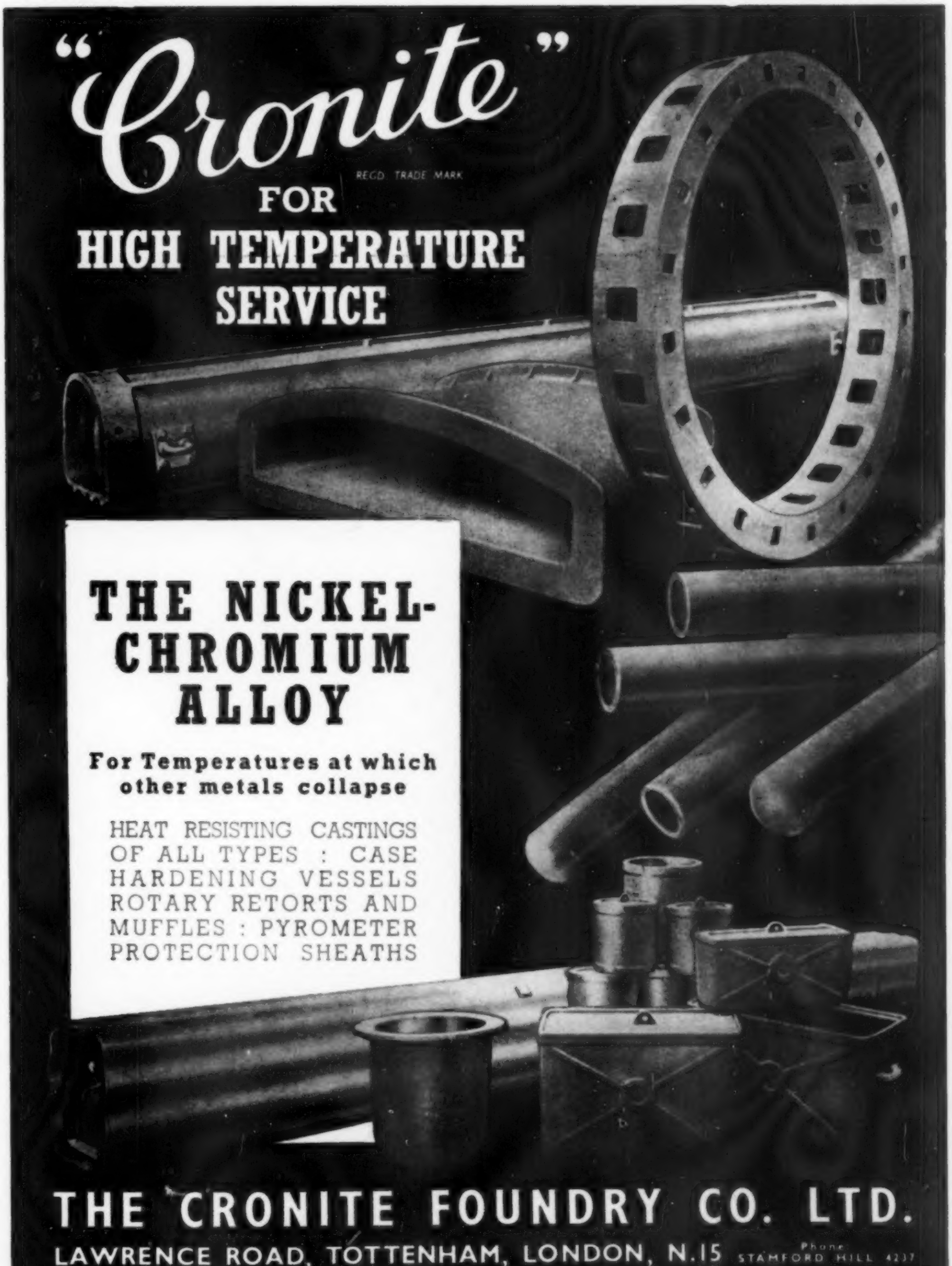
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